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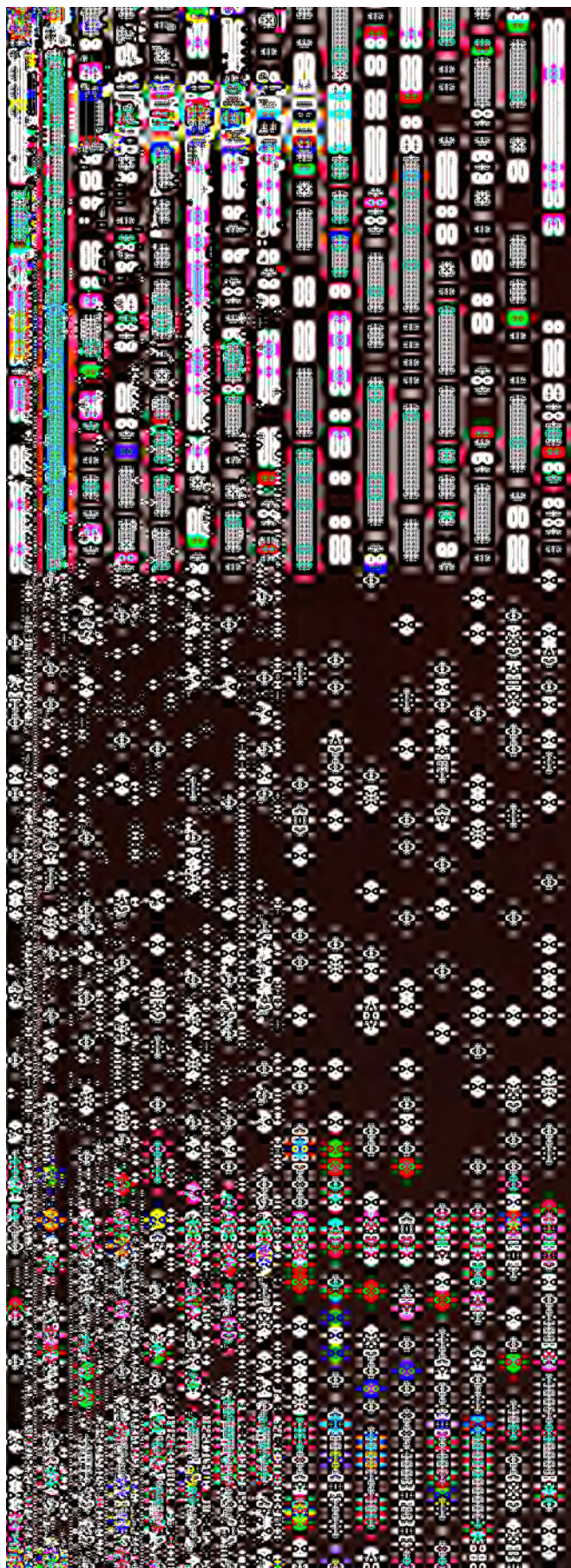
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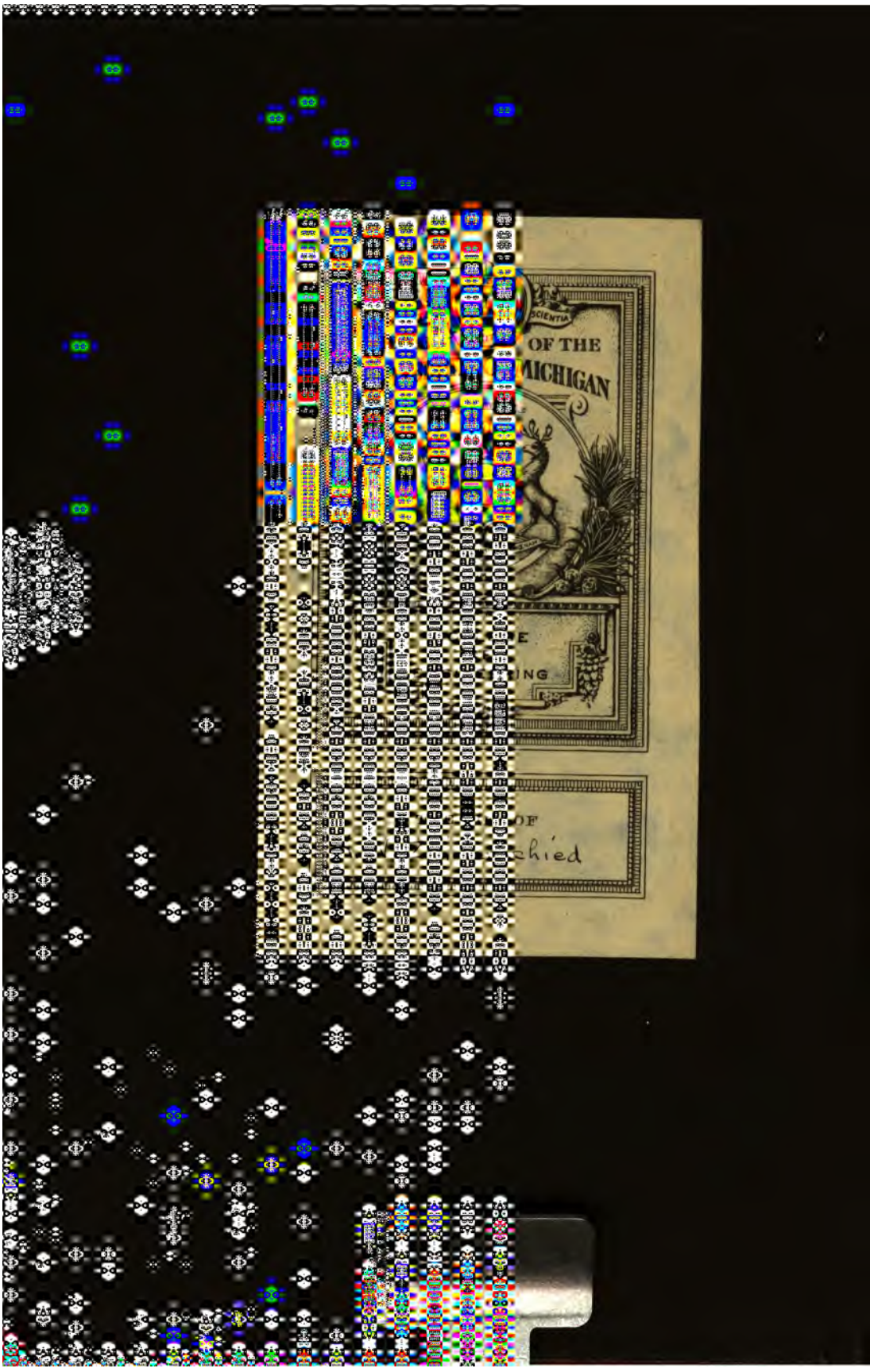
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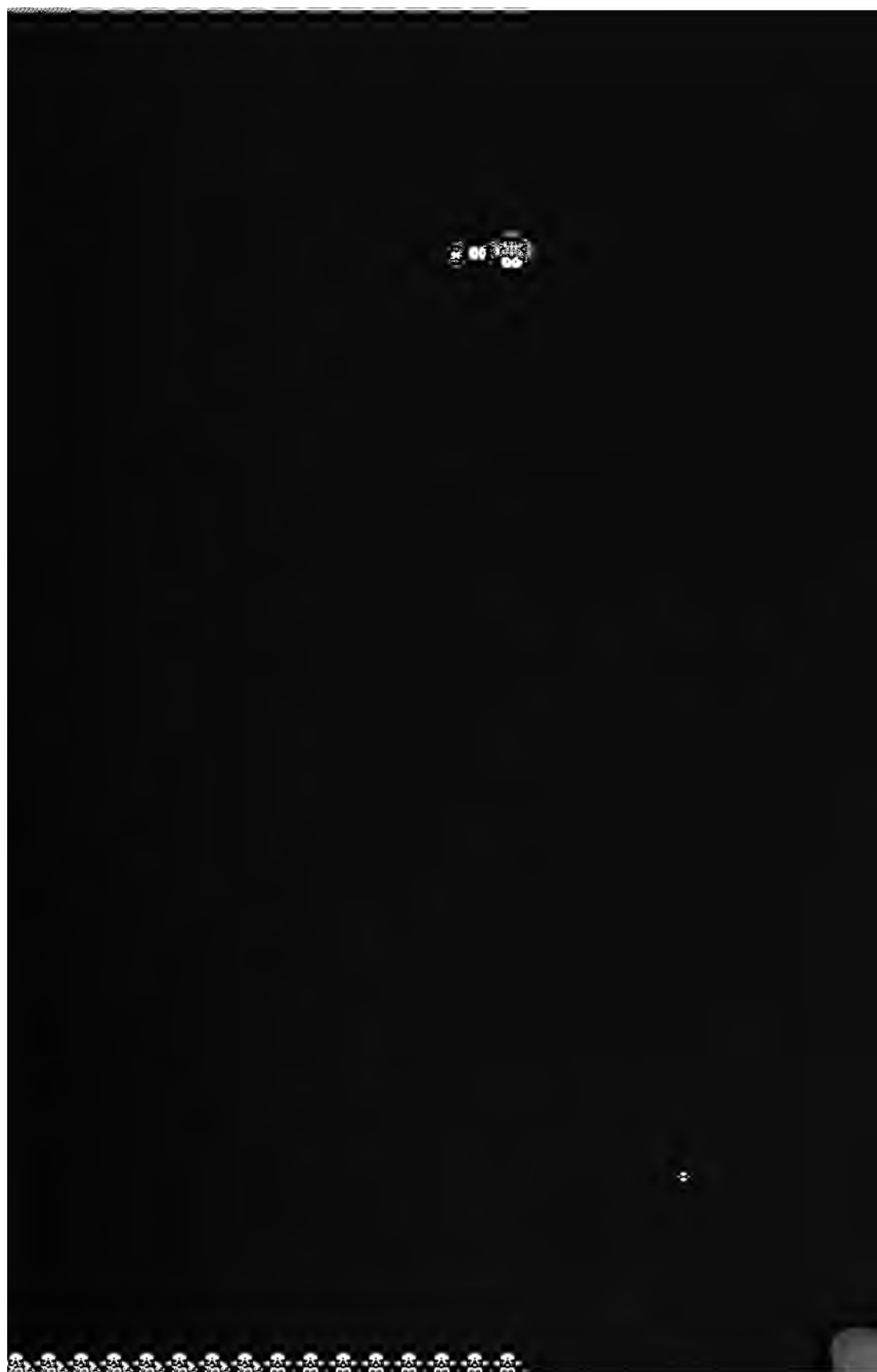
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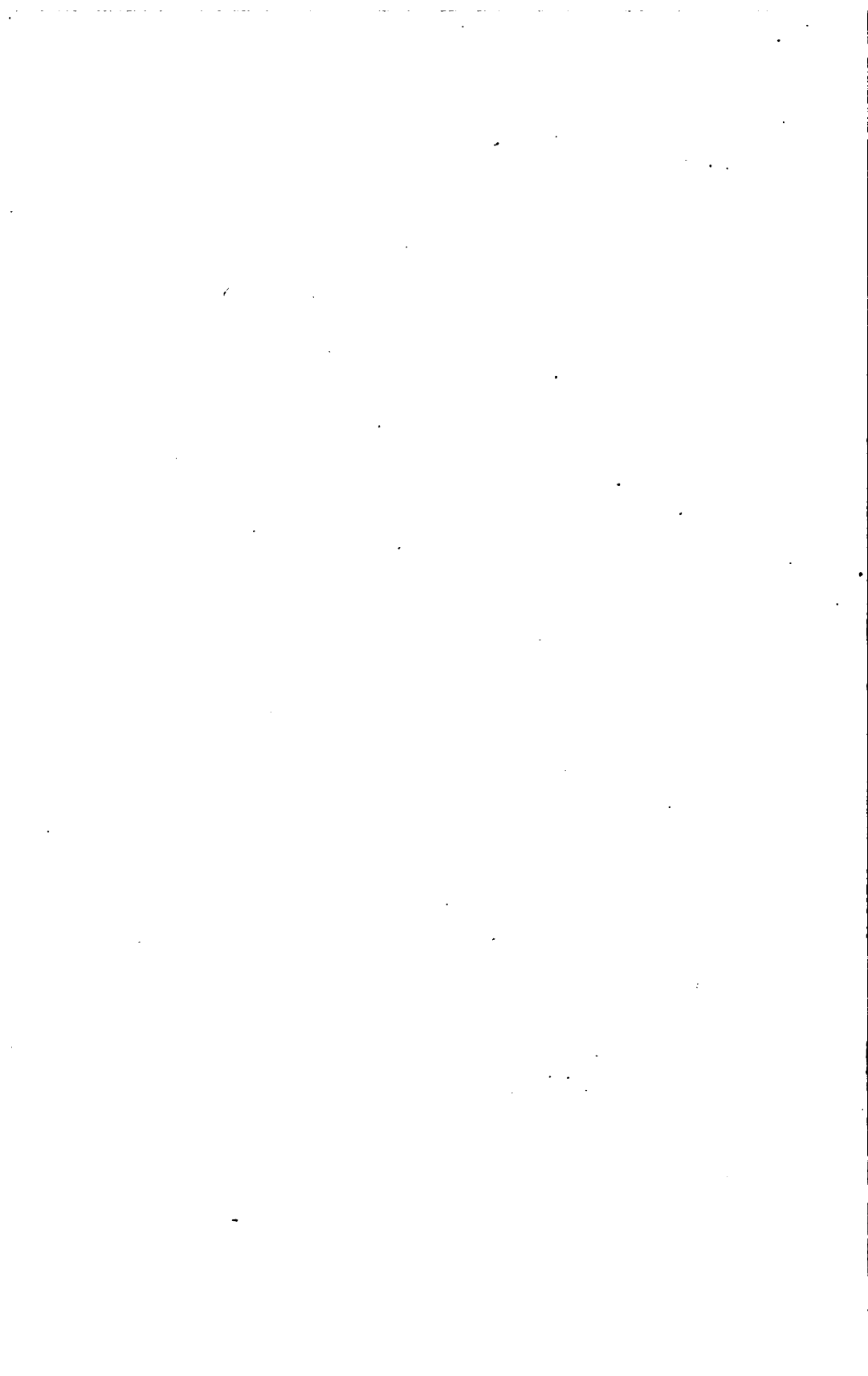












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The Structure and Properties  
of the more common  
Materials of Construction

BY  
G. B. UPTON


*Assistant Professor of Experimental Engineering  
Sibley College, Cornell University*

FIRST EDITION

FIRST THOUSAND

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## PREFACE

THIS book is an outgrowth of the laboratory course in Materials of Construction given to Juniors in Sibley College. For some time past the instructors in the course have been giving, during the first hour of each three hour laboratory period, a lecture upon the theoretical side of the particular test being conducted with indications of the relation of the test to engineering practice. Then came the actual testing work in the laboratory as a demonstration of the theory, with further discussion during the conduct of the test. Both the theory and conduct of the test and its relation to engineering practice were then discussed in the reports written by the students. In time it has become evident that we cannot by this method thoroughly cover as much ground as we desire. Hereafter students will study the theory of each topic in this book and discuss theory and practice with the instructor during the first part of the laboratory period. This discussion will replace the lecture formerly given.

The book is restricted almost entirely to the theoretical discussion of the field. It is divided into two parts. The first part deals with the determination of the properties of the materials by means of engineering testing. An analytical attitude, mainly from the standpoint of metallography, has been used throughout. Under each method of loading as tension, torsion, etc., there is a discussion of what goes on inside the material throughout the progress of the loading to break. The ordinary formulas for calculation of stresses and deformations are examined. The assumptions on which they are based are stated. Their range of validity is given and the direction and magnitude of their errors after they cease to be valid. Knowledge of these points is considered essential to the intelligent use of the formulas. The factors changing the results of tests are discussed, bringing out the need of standard specifications. No attempt is made to give collections of data of results of tests, but there is an attempt to teach a student how to interpret and criticize test results.



The second part of the book deals with the nature of the internal structure of materials and the control of properties through the control of the internal structure. This part of the book becomes a short study of theoretical and applied metallography. Realizing, however, that the scientific language of metallography might scare the student into thinking the subject more difficult and complicated than it is, care has been used to present the principles of physical chemistry concerned in simple words, and with many diagrams to make relations clear.

Certain parts of the book are, so far as the author is aware, either entirely new or have never before been given in similar detail. Such are the general analysis for finding of true stresses after the yield point in torsion and transverse loading; and the detailed description and explanation of heat treatment from the standpoint of physical chemistry. The new method for torsion and transverse test analysis may clear up the controversies concerning "Guest's law" or other laws for conditions of failure in combined loadings, and may make the tests of ductile metals in direct shear unnecessary. The new analysis of the general nature and of the detail of heat treatment, presented mathematically and in diagrams, already throws much light upon the problems, and may be the beginning of a complete theory.

Particular thanks for assistance in the composition of the book, criticism of the text, etc., are due to Mr. W. R. Wigley and Mr. W. J. Diederichs.

G. B. UPTON.

SIBLEY COLLEGE, ITHACA, N. Y.

*October, 1915.*

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# MATERIALS OF CONSTRUCTION

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## CHAPTER I

### DEFINITION AND EXPLANATION OF TERMS

1. **Loads** are the external forces applied to a body. Loads are of various kinds, according to their method of application to the piece. We distinguish (1) **tension** loading, tending to pull the piece in two; (2) **compression** loading, pushing on the piece; (3) **transverse** loading, tending to bend the piece; and (4) **torsion** loading, twisting the piece. Material in practice is frequently under two or more kinds of loading simultaneously, as tension and transverse, or transverse and torsion.

2. The effect of a load upon a piece of material is to change the shape of the piece. This change of shape is called a **deformation**. Each kind of loading has its appropriate deformation. For tensions, the piece lengthens parallel to the load and becomes thinner perpendicular to it. Compression reverses the tension results. Bending of the piece is the external, observable deformation in transverse loading; and twisting is the external deformation in torsion loading. Under proper commercial loadings the deformations of pieces are negligibly small; generally they are too small to be detected save with delicate instruments.

3. Material resists the deforming action of loads. The resistance comes from internal forces distributed within the material. These distributed internal forces, by which a body resists deformation due to external loads, are called **stresses**. Or, stresses are the internal forces by which one part of a body transmits loads to adjacent parts.

4. No matter what the kind of the external loading, the internal

stresses set up in a piece are of two general kinds only: **normal** stresses and **tangential** stresses. In Fig. 1, let  $A$  be a piece under load. (The loads are not indicated.) Imagine the piece to be cut on the plane  $xy$  and the part above  $xy$  removed. The stresses by which the part above was acting on the part below  $xy$  may be represented either by forces

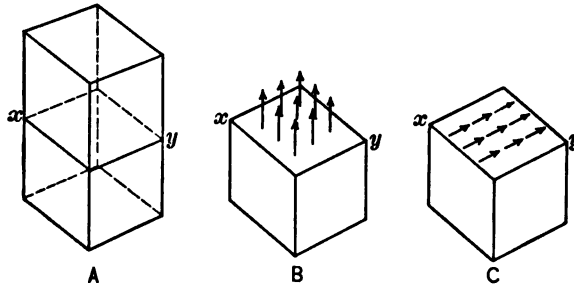


FIG. 1. Stresses.

perpendicular to  $xy$ , as in  $B$ , or parallel to  $xy$ , as in  $C$ , or by a combination of the two. There is no possible loading of the piece in which the stresses acting across  $xy$  cannot be resolved into sets of stresses normal to, and parallel to,  $xy$ .

Normal stresses are of two kinds: tension and compression.  $B$  in Fig. 1 shows tension stress. The difference between tension and compression stresses is of the algebraic sign only; if one be taken as positive the other is negative.

Tangential stresses, forces parallel to an imaginary exposed section in the material, are called **shears**.

5. In general, both normal stresses and shears are to be expected in a piece of material under load. It is often possible, as will be shown later, to find planes through a piece on which there are normal stresses only, or shears only; but on an analysis plane placed haphazard both normal stresses and shears must be accounted for.

6. Each kind of stress is accompanied by its appropriate deformation. In Fig. 2, representing a side of a cube, the general nature of the deformation is indicated. The normal stresses cause mainly change of length; shears cause twist of a body or slip of one surface over another.

In Fig. 2,  $A$  and  $B$ , tension and compression, the deformation

usually measured is the change of length,  $\Delta L$ . Imagine a pile of cubes, such as *B*, placed one above another, and the same load transmitted through the lot. Then each will change its length by  $\Delta L$ . If there are  $n$  cubes, the entire assembly will change its length from  $nL$  to  $n(L + \Delta L)$ , or the total change will be  $n(\Delta L)$ . The relative change will be  $(n\Delta L) \div (nL) = \frac{\Delta L}{L}$ . It is obvious then that by calculating the **relative deformation** we have found a quantity which is

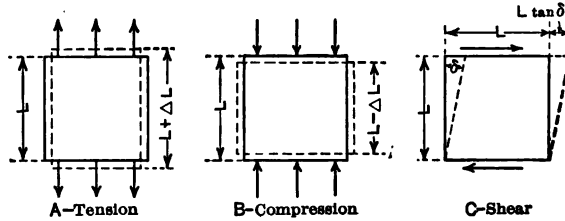


FIG. 2. Deformations.

independent of the absolute length of the body, and which is characteristic of its material and loading. Incidentally this relative or **unit deformation** is an abstract quantity or number, a ratio, for it is (change of length)  $\div$  (length); and hence, by a shift of decimal point, relative deformation may be read as a percentage change of shape. Example:  $\frac{\Delta L}{L} = 0.10$  means a 10 per cent change of length.

By similar reasoning to that of the preceding paragraph it will be seen that for shears the relative deformation is  $L \tan \delta \div L$ , or  $\tan \delta$ . This likewise is an abstract number characteristic of the material and its loading, and independent of the gross dimensions of the body.

7. The measure of stress is **stress intensity**. The external load is distributed among, and met by, the stresses acting over an area of section of the piece. Stress intensity, or **unit stress**, is of the dimensions force  $\div$  area; in American practice, pounds per sq. in. For example, a bar of rectangular section, 2 inches by 1 inch, in a bridge, carries a tension load of 25,000 pounds. The intensity of the tension stress in the bar is then  $25,000 \div (2 \times 1) = 12,500$  pounds per square inch. If the bar were 5 inches by 1 inch cross-section, it would have more metal to carry the same load; or with the same load the metal



would be stressed to only  $25,000 \div (5 \times 1) = 5000$  pounds per square inch.

8. The **strength** of a material is the stress intensity, pounds per square inch, which will cause the material to fail in some definite way.

Two pieces of different sizes and of the same material are similarly loaded in proportion to their strengths when the stress intensities in the two pieces are identical.

9. We have seen above that the gross dimensions of the body are eliminated in deformation measurement by calculating the relative or unit deformation. Now we find that the gross dimensions of the body are eliminated from the load measurement by reducing to stress intensity, or unit stress. Hence, if we consider stress intensity and relative deformation as functions of each other, we must find that the relation between them is characteristic of the material, for other considerations have been eliminated.

10. The curve of stress intensity against relative deformation, in its most general form as found in engineering materials, is shown in Fig. 3. Starting from zero there is first, along *AB*, a range of loading

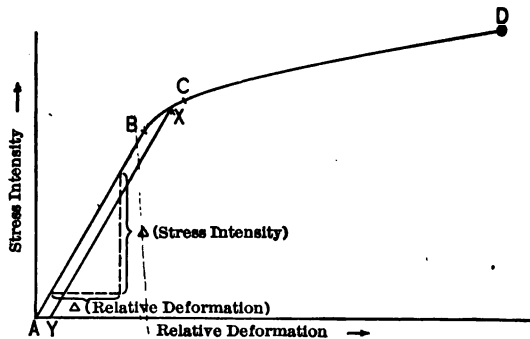


FIG. 3. General Stress-deformation Curve.

for which stress intensity and deformation are closely proportional to each other. On unloading from any point under *B* it will be found that the unloading line of the stress-deformation relation practically coincides with the loading line, and that when all the load has been removed the piece has also recovered all its deformation, and has returned to the condition represented by the starting point *A*. Such action is called **elastic action**. *AB* is called the **elastic line** for the

material. The strict criterion of elastic action is the perfect return to the point *A*. More common, but less exact, definitions of elastic action are based upon the proportionality of stress intensity and relative deformation on the elastic line, and on the coincidence of loading and unloading lines.

The point *B*, where proportionality between stress intensity and deformation ceases, is called the **elastic limit**. After passing this point the material cannot of itself return to the point *A*. Unloading from point *X*, for example, will go, not along *XBA*, in reverse of what happened in loading, but along *XY*, practically parallel to the elastic line *AB*. The deformation *AY* will not be recovered. It is a **permanent deformation**, or **permanent set**.

On reloading, the material will closely retrace *YX* to *X*; then further loading will go out *XCD*, practically as if the unloading and reloading *XYX* had never taken place. Such a line as *XY*, of unloading and reloading to show the amount of permanent set, is called a **set line**. Set lines are nearly, but not exactly, parallel to the elastic line in most engineering materials. Such set lines as *XY* may be taken not only with *X* between *B* and *C*, but with *X* any point on the main stress-deformation curve beyond *B*, clear to the break point *D*.

From *B* to *C*, increasing load has been causing increasing, but small, permanent deformation. At the point *C* the deformation rate markedly increases, so that beyond *C* even small increases of load cause large increases of permanent deformation. Hence *C* is called the **yield point**. Obviously, a piece of material will lose its usefulness as a part of an engineering structure if while in service it becomes stressed beyond the yield point, for the consequent large change of shape will spoil its fit and relation to neighboring parts. The smaller permanent deformations between the elastic limit and the yield point are often negligible in practice. Hence in engineering specifications the **yield point stress intensity is an important property of a material**. By looseness of usage in commercial language the yield point is generally spoken of as the "elastic limit," the true elastic limit being disregarded. This confusion of elastic limit with yield point in commercial usage is due to the fact that in worked steels, the most common of structural materials, the two points are very close together.

The shape of the stress-deformation curve from *C* to *D* (yield point to break) varies considerably both with the material and with the method of calculation of stress intensity and relative deformation. More definite discussion of this will be given in the chapter on tension loading of ductile metals.

11. The general character of elastic deformations has been indicated in Fig. 2 and its discussion. In kind, the elastic deformations for tension or compression are like the familiar actions of rubber, while the elastic shear deformations may be studied from a cube of jelly. In size, the elastic deformations in engineering materials are generally very small — too small usually to be visible to the naked eye. Permanent deformations are different in nature from elastic deformations. It can be shown experimentally that the density and volume of a body do not appreciably change while the body undergoes even the largest permanent deformations. There is but one conceivable mechanical explanation for permanent deformation: the internal structure of the body is rearranged during the deformation, by parts slipping on adjacent parts. Such slipping of one part over another is a shear deformation. Hence we deduce the important principle that the normal stresses, **tension and compression cannot of themselves cause or account for any permanent deformation; such a deformation can be caused only by a shear stress.** Break may be caused by either tension or shear.

12. The scheme of the permanent deformation is shown in Fig. 4. Fig. 4 *A* represents a piece of material made of cubic parts fastened one on another, without load or deformation. *B* shows the result of loading within the elastic limit. The body as a whole and each of its parts have twists measured by  $\tan \delta$ . *C* shows the effect of carrying the load beyond the elastic limit. The deformation of the whole body is now  $\tan (\delta + \delta')$ . The elastic deformation of each part is  $\tan \delta$ , larger than in *B* in proportion as the load is larger. The difference between  $\tan (\delta + \delta')$  and  $\tan \delta$  is the permanent deformation, built up of the successive slips of the parts over each other. Lastly, *D* shows the condition of the body after the loading of *C* has been removed. Each part recovers from its elastic deformation; the slips remain. The actual mechanism of permanent deformation in a solid body is of course far more complicated than the

scheme of Fig. 4; but that this scheme is correct in principle will be shown in Chapter II, in the description of the actual internal structure of materials and its action under load.

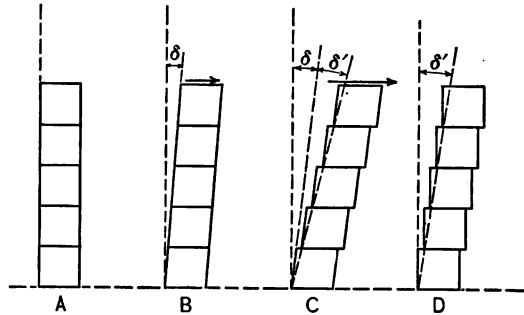


FIG. 4. Mechanical Scheme of Permanent Deformation.

13. It can readily be shown that no matter what the external loading may be there are planes in the material along which shear slipping tends to occur. Since all permanent deformation consists of such slipping, and since its amount depends on the intensity of the shear stress, it is obviously important to know for every case of loading the direction and magnitude of the maximum shear stress. The appropriate mathematical formulas will be developed in the special chapters on the various kinds of loading.

14. There are two quantities determined by the relation of stress intensity and relative deformation which are of importance to the engineer. The first is the **stiffness** of the material. The stiffness is measured by the ratio of stress intensity to relative deformation along the elastic line, or during elastic action; more accurately, it is the slope of any line of elastic action. For, the less the deformation caused by a given stress intensity, or the greater the stress intensity required to cause a given deformation, the stiffer is the material; and it is not fair to the material to load it save elastically while determining its elastic stiffness. The particular name given to the ratio  $\Delta$  (stress intensity)  $\div$   $\Delta$  (relative deformation), along a line of elastic action, as a measure of stiffness, is **modulus of elasticity**. See Fig. 3, where  $\Delta$  (stress intensity)  $\div$   $\Delta$  (relative deformation) =  $E$ , the modulus of elasticity.

There are three moduli of elasticity: (1) the **tension-compression modulus** (the modulus for compression is theoretically and practically the same as in tension); (2) the **shear modulus**; (3) the **volume or bulk modulus**. When one speaks of "modulus of elasticity" without further specification, the tension-compression modulus is meant. In physics it is often called Young's modulus. It is measured by (normal stress intensity)  $\div$  (relative change of length). The shear modulus is also called the **modulus of rigidity**. It is measured by (shear stress intensity)  $\div$  (relative twist, or  $\tan \delta$  value). The bulk modulus has little application in engineering. It is measured by (hydrostatic pressure)  $\div$  (relative change of volume). It can be shown that these moduli are so interconnected that any two of them determine the third.\*

15. The other secondary quantity determined by the relation of stress intensity to relative deformation is a measure of the energy absorbed by the material, acting as a spring, while being loaded. This energy, if the loading is elastic, is recovered from the material during unloading. It is called the **resilience** of a material; the **modulus of resilience** of a material is the resilience in inch-pounds of work per cubic inch (unit volume) of material, for a loading which goes to the limit of elastic action — that is, to the elastic limit. Let  $AB$ , Fig. 5, represent the elastic line of loading for a given material; then  $CB$  is the stress intensity at the elastic limit, and  $AC$  is the elastic relative deformation at the same point. Work is the product of force by distance. By dividing the triangle  $ABC$  into vertical strips, each of a height = stress intensity, and of width =  $\Delta$  (relative deformation), it is evident that the area of each strip represents

\* If the material is isotropic, that is, has the same properties in all directions, and if  $E_t$  stands for the tension modulus,  $E_s$  for the shear modulus, and  $E_v$  for the bulk modulus, then

$$E_v = \frac{E_s E_t}{9 E_s - 3 E_t}.$$

Also, let  $\mu$  = Poisson's ratio = the ratio of relative lateral (linear) constriction to relative change of length by elastic stressing in tension or compression; then

$$\mu = \frac{E_t}{2 E_s} - 1 = \frac{3 E_v - 2 E_s}{6 E_v + 2 E_s}.$$

$E_t$  and  $E_s$  are easily found by experiment;  $E_v$  and  $\mu$  may then be computed. Direct measurement of  $E_v$  or  $\mu$  is difficult.

$\Delta$  (work done), and the sum of all the strips, or the area of the triangle  $ABC$ , is really the measure of the elastic work or resilience.

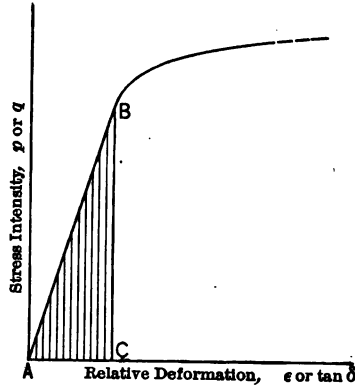


FIG. 5.

Let stress intensities and relative deformations at the elastic limit be represented by

- $p_E$  = tension or compression stress intensity,
- $q_E$  = shear stress intensity,
- $\epsilon_E$  = tension or compression relative deformation,
- $\tan \delta_E$  = shear relative deformation.

Then the maximum elastic resilience, or modulus of resilience,

$$U = \frac{p_E \epsilon_E}{2} \text{ OR } \frac{q_E \tan \delta_E}{2},$$

as the case may be. By engineering convention, however, when making the calculation of modulus of resilience, we substitute for the elastic limit point either (1) the yield point, in ductile materials, or (2) the break point, in brittle materials, because to these points the action of the material is sufficiently near to elastic to answer the engineering requirements, and these points are much better defined than the true elastic limit points.

## CHAPTER II

### GENERAL NATURE OF THE INTERNAL STRUCTURE OF MATERIALS AND ACTION UNDER LOAD

16. Engineering materials of construction may be classified according to their structures under three headings: first, wood; second, stone, brick, cement, mortars, concrete; third, metal.

Wood is built up of systems of parallel overlapping fibers. Although these fibers cannot in any sense properly be said to be of crystalline structure, yet the effect of the fibrous structure upon the properties is precisely what would occur if the fibers were elongated crystals.

All solid matter is either **crystalline** or **amorphous**. The ultimate physical unit in matter is the molecule. When the arrangement of molecules in a particle of matter is according to a definite repeating geometrical pattern, the particle is a crystal. When all definite geometrical arrangement is lacking, the particle is amorphous.

Some building stones are crystalline aggregates. Granite is an example; the crystals show plainly to the naked eye on a polished granite surface. More usually, building stones are like concrete, a mixture of crystalline and amorphous material.

17. Examination under the microscope of the internal structure of any of the engineering metals shows that the metal is an aggregate of an immense number of minute crystals. The separate crystals in a steel, for example, vary from a maximum diameter of 0.01 or 0.02 inch down to around 0.0001 inch and even smaller, sometimes being too small even for the highest powered microscope to make them visible.

The external shapes of these crystals are irregular, being determined simply by their interferences with their neighbors. At their bounding surfaces the crystals are firmly and strongly interlocked with each other; fractures of the metal always (in engineering metals) go through the crystals, not between them along the boundaries, merely pulling the crystals apart.

The break of the metal, then, whether it occurs at the end of a single



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break of the metal  
the separate crystals

minute to be studied  
individually, the unit  
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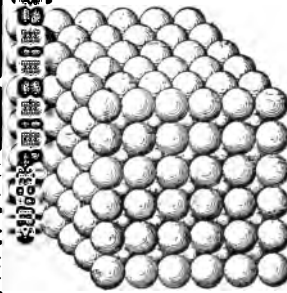


FIG. 6.

shot. Fig. 6 shows  
system of crystalliza-  
forms, the crystals of

determined by the nature  
in which they are put  
are as shown in Fig. 6  
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# STRUCTURE

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of maximum intensity of shear  
formation of amorphous material  
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it is easier to start a new slip  
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ncreases its resistance; a third slip  
s, and so on, until the crystal has  
the stress by slipping an in-  
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le slips are all permanent, per-  
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these systems of parallel slip  
revealed by the microscope as  
of lines marking the polished  
through the crystals.  
account of the geometric pat-  
arrangement of molecules in  
crystals there are usually certain planes  
crystal along which it is rela-  
crystal structure. These are  
planes of the crystal. Material  
action of these slip planes

while the material is being loaded. The classes are (1) **brittle**, (2) **ductile**.

22. A brittle material is one which breaks without taking any appreciable permanent deformation. A ductile material takes an appreciable permanent deformation before breaking. In a brittle material the slightest slip through a crystal generally becomes a permanent break of that crystal. The parts of the crystal do not stick together across a slip plane after slip starts. In ductile material the parts of the crystal do stick together after slip starts, while resistance to continued slip increases with the slip. Hence a slip along any given plane is generally very minute, being stopped by the beginning of slip along an adjacent parallel plane. By the accumulation of large numbers of small slips, the crystal as a whole may show a considerable change of shape before it breaks. Break in a brittle metal may, but does not usually, follow crystal boundaries; in a ductile metal break is always through the bodies of the crystals, not along their boundaries.

Fig. 7 shows how these slip planes look in a ductile material. The surface of a bar of ductile metal was polished, and the bar then loaded (tension) beyond its yield point. A photomicrograph was then made of the surface, using oblique light. The light coming from the side throws on the surface shadows of the edges of the slips. The slips show in the series of parallel lines across the crystals. It will be noted that there are three intersecting systems of slip planes through these crystals. Remembering that the crystals themselves have to be highly magnified to make them visible and seeing here that there are hundreds of slip planes through a single crystal, the scale of magnitude of the slips may be realized.

## CHAPTER III

### TENSION LOADING OF BRITTLE MATERIALS

23. By far the most common brittle metal in engineering practice is "gray cast iron." This is an excellent material to take for the study of the tension loading of brittle materials because it is very complex in structure and therefore will bring in all the probable and possible phenomena of the case.

24. The chemical analysis, in percentage by weight, of a possible gray cast iron is:

	Per cent
Chemically combined carbon.....	0.6
Chemically combined phosphorus.....	0.5
Chemically combined sulphur.....	0.1
Free carbon or graphite.....	3.0
Silicon.....	2.5
Iron.....	93.3

The combined carbon is in the form of  $\text{Fe}_3\text{C}$ , "cementite," a strong, brittle substance, somewhat harder than glass. Similar to this physically are the phosphorus compound  $\text{Fe}_3\text{P}$  and the sulphur compound  $\text{FeS}$ . The **graphite** is a soft black crystalline substance, of physical properties quite similar to chalk. The silicon is partly in the form of  $\text{Fe}_3\text{Si}$  crystallized with  $\text{Fe}_3\text{C}$ , and partly crystallized with the iron in the crystals of "ferrite." Ferrite is the name given to the ordinary form of iron; physically it is a soft, rather weak, very ductile and peculiarly magnetic metal. **Silico-ferrite**, ferrite with silicon crystallized in it, is less ductile than pure ferrite.

25. We may now rearrange our analysis into a physical analysis of parts of similar properties.

For discussion of the effect of the different parts on the physical properties of the whole, we must use the volume analysis below. Here the graphite greatly increases in importance, since it is of far lower specific gravity than the other parts. For the present we may con-

sider the gray cast iron to be a uniform and intimate physical mixture of the parts given above. The size of the separate crystals is just at, or somewhat below, the limit of visibility with the naked eye. Crystal groups, making the "grain" in cast iron, are visible to the naked eye.

Physical analysis	Per cent by weight	Per cent by volume
Hard, brittle, strong..... $\left. \begin{array}{l} \text{Fe}_3\text{C} \\ \text{Fe}_3\text{Si} \\ \text{Fe}_3\text{P} \\ \text{FeS} \end{array} \right\}$	17	16
Soft, brittle, weak.....Graphite	3	9
Soft, ductile, low yield point, but strong at break, silico-ferrite....Fe(Si)	80	75

26. Consider now the effect of an increasing tension load upon the internal structure of a piece of gray cast iron. At the beginning of loading all of the crystals will deform equally. Actual stress intensities in the different crystals will vary with the nature of the crystal, since the stress intensity depends upon both the deformation and the modulus of elasticity. When, therefore, we calculate the stress intensity by dividing the total tension load by the area of the cross-section of the piece, we get, not the real stress intensity in any one crystal, but an average value for the piece. **We never really do know the actual stress intensity in any given crystal.** As the loading progresses, the graphite crystals, weakest of all those present, are broken, beginning with those least favorably situated for resisting the load. These graphite crystals cannot fall apart after they are cracked and broken because they are bedded in a surrounding structure which is largely composed of the ductile ferrite. Actual stress intensity in the graphite crystals cannot appreciably rise above the value which first breaks them.

Any further increase in the loading must be carried by other crystals than those of graphite. At this stage or before, some of the ferrite crystals have passed their yield points; they begin to flow in a semi-plastic manner, by formation of slip planes. Henceforth we realize more nearly stress equality than deformation equality from crystal to crystal. The amount of the hard crystals is so great, and their

disposition with regard to the ferrite is such, that the gray cast iron as a whole cannot show much of permanent deformation before reaching the breaking point of the brittle crystals, and therefore of the entire piece.

27. From this study of the internal action, the form of the stress-deformation curve *ACE* of the gray cast iron, Fig. 8, will be under-

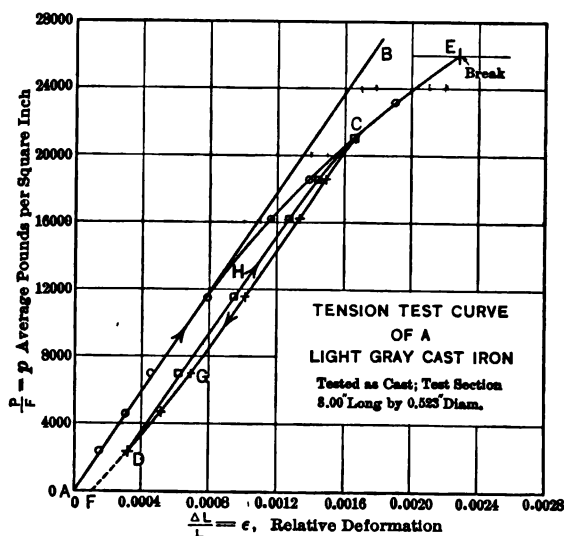


FIG. 8.

stood. From almost the beginning of loading the material begins to take slight permanent set. As the average stress increases the rate of permanent set increases also.

The curve of first loading of a cast iron shows no true elastic limit and in general is a curved line, becoming flatter and flatter with increase of load. True elastic action of the cast iron would be indicated by the straight line, *AB*, tangent to the curve at the origin. (The yield point of the ferrite may appear as a knee of the curve, at about 20,000 pounds per square inch, in light gray cast irons and "semi-steels." On casual inspection, or with measurements lacking in precision, this knee of the curve may be mistaken for an elastic limit.)

If the loading be gradually increased until break comes, the curve

of action will be  $ACE$ . If from any point  $C$  the load is removed, the unloading curve is not  $CA$  but  $CGD$ . Reloading will follow  $DHC$ . Beyond  $C$  the loading will continue on the main curve as if the unloading had not taken place, **provided the unloading and reloading has been done only a few times.** The line  $CD$  is called a **set line**;  $AF$  is the permanent set caused by loading as far as  $C$ . (The unloading line of  $CGD$  has been produced to  $F$ . In the test plotted in Fig. 8 the load was not actually taken below  $D$ .)

28. Attention must now be given more closely to the peculiar form of the set lines. **They are not straight lines; the "down" line lies below the "up" line, so that a loop is enclosed.** This loop is truly a hysteresis loop of stress-deformation, due to internal friction of the material. The beginning of unloading, and the beginning of reloading, are elastic lines, parallel to  $AB$ ; for the rest of the loop the lines are flatter than elastic lines, and combine permanent deformation with the elastic.

In the material as it exists at the point  $C$  some crystals have previously broken down and are carrying less than the average stress intensity, with permanent deformation; other crystals are still elastic, and have stress intensities considerably above the average. When the external loading is removed we get at first elastic reduction of stress and deformation throughout all of the crystals, along  $CG$ . Then those crystals which had taken permanent set, and which had stress intensities less than the average, get to zero stress, although the average stress on all the crystals is still of considerable value. From this point those crystals which were overly stressed, but still elastic at the point  $C$ , act as springs, **reversing the stress and beginning to reverse the permanent deformation of the weaker crystals.** When the external load and the average computed stress reach zero at the bottom of a set line, **the crystals have not really been unloaded if any permanent set remains,** but are partly in tension, partly in compression. This condition of internal stress from crystal to crystal, after it has once been set up, can be relieved only by heat treatment of the metal. **Removal of the external load does not relieve the internal stresses, if there is permanent set.**

The farther we go out the main curve before taking a set loop, the larger and wider is the loop.

29. If we unload and reload repeatedly, using a given set loop as a cycle, we may ultimately break the material without ever carrying on the loading beyond the upper end of the loop. The larger the loop the fewer are the repetitions of it required for the break. Further discussion of this action belongs in the chapter on fatigue failure.

30. For engineering purposes it is customary to make a number of approximations as to the properties of cast iron. For instance, up to about 10,000 pounds per square inch the permanent deformation is only about 0.0001 inch per inch, or less, an amount usually imperceptible in the engineering use of a metal. We might, indeed, arbitrarily define the engineering elastic limit value as that stress which first shows a permanent set as large as 0.0001 relative deformation. Another arbitrary definition of the engineering elastic limit is that of J. B. Johnson — that point of the main curve at which its slope is two-thirds of the slope at the origin. For calculating the engineering value of the modulus of resilience of cast iron we assume the elastic limit at the breaking load, and also that the curve *ACE* is a straight line; then  $\frac{p_B \epsilon_B}{2}$  is the modulus of resilience. (Subscript *B* refers to

break.) For modulus of elasticity, engineering practice usually takes the average slope of the curve up to the arbitrary engineering elastic limit, instead of the true value, the slope of *AB*, Fig. 8. There is apparently no consistency in the engineering conventions above; yet each convention will be found to have some reason for its existence when we consider the practical use made of the different quantities.

31. In order to get a fair and satisfactory test certain precautions must be used as to the form of the test piece and the method of holding it. The ends of the test piece, which are gripped in the heads of the machine, are under a double loading, carrying both the lengthwise tension pull and a sidewise compression from the grips. Hence to avoid a premature failure by break in the grips the ends of the test piece must be larger than the test section. The cross-section area in the grips should be about twice the area of the test portion of the piece. The change of section area from the grip section to the test section should be gradual. Corners should be rounded. Fig. 9A is a good form, Fig. 9B a bad form, in the way the section change



should take place. The objection to 9B is that it gives an uneven and non-parallel distribution of tension stress across the section at  $x$  and  $y$ , as indicated in Fig. 10. The maximum tension stress at the outside of the section at  $xy$  is larger than the average stress at  $uv$  in



FIG. 9A.

the test portion proper of the test piece, and causes a premature break at  $xy$ .

32. The most common cause of error in handling the test is eccentric loading. If the resultant total pull on the test section from the

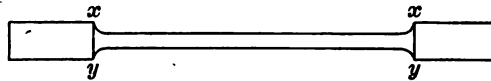


FIG. 9B.

grips does not pass through the center of the test section, the loading is eccentric. The result in stress distribution through the section



FIG. 10.



FIG. 11A.



FIG. 11B.

is shown in Fig. 11A and B. Instead of having uniform stress  $p = \frac{P}{F}$  ( $P$  = load;  $F$  = area of cross-section), with eccentric loading we have a variable stress, with a maximum stress intensity much larger than the average. This maximum stress determines the break, which comes prematurely, at a lower value of load than should have been

obtained. In a round test section, if  $d$  is the diameter of the section,  $x$  the eccentricity of the load,

$$p_{\text{maximum}} = (p_{\text{average}}) \left( 1 + \frac{8x}{d} \right).^*$$

The eccentricity is determined largely by the method of gripping the ends of the test piece.

33. There are two ways in common use of holding the ends of a test piece in the heads of the testing machine — wedge grips and screw grips. The wedge grips are shown in Fig. 12. In the upper head  $AA$  of the testing machine is a wedge-shaped slot. Into this slot fit the wedges  $BB$ , holding between them the test piece  $C$ . The

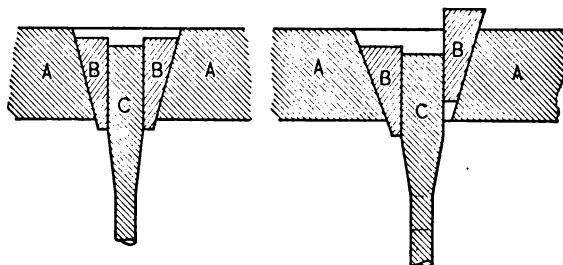


FIG. 12A.

FIG. 12B.

wedges are grooved and roughened on the side toward the test piece, but smooth to the testing machine. Hence a pull on  $C$  draws the wedges with the test piece down into the slot, and tightens the grips as the load increases. The trouble with the wedge grip is that one wedge may slip while the other holds, resulting in a situation indicated

\* From direct tension  $p = \frac{P}{F}$ . From bending,  $p = (\text{bending moment}) \times (\text{radius of section}) \div (\text{moment of inertia of section})$ . Bending moment = (load)  $\times$  (eccentricity). Hence

$$p_{\text{bending}} = (Px) \left( \frac{d}{2} \right) \div \left( \frac{\pi d^4}{64} \right) = \frac{32 x P}{\pi d^3} = \frac{8 x P}{d \cdot \left( \frac{\pi}{4} d^2 \right)}.$$

But  $\frac{\pi}{4} d^2 = F. \quad \therefore \quad p_{\text{bending}} = \frac{P}{F} \cdot \frac{8x}{d}.$

The maximum stress in the section is the sum of the direct tension and the bending tension stresses, or

$$p_{\text{maximum}} = \frac{P}{F} + \frac{P}{F} \cdot \frac{8x}{d} = \frac{P}{F} \left( 1 + \frac{8x}{d} \right) = (p_{\text{average}}) \left( 1 + \frac{8x}{d} \right).$$

in an exaggerated way in Fig. 12B — a badly eccentric loading. Sometimes the two wedges are so linked that they must move up and down together. Even then one may hold the test piece while the other slips, so that eccentric loading is apparently, but not really, avoided.

For the screw grips the ends of the test piece are threaded. The threads are made loose fits for nuts suspended through universal joints from the heads of the testing machine. This combination of screw grip and universal joints ensures very nearly concentric loading, with freedom from lateral constraints. It should be required in all testing of any pretension to scientific accuracy.

34. A break conditioned by a tension stress is practically instantaneous when the maximum tension stress intensity reaches the tension strength of the material.\* The direction of the break is perpendicular to the direction of the maximum tension stress acting. Let Fig. 13 represent a part of the test section of a tension test piece, concentrically loaded by a force  $P$ . The normal cross-section is  $F$ ; at an angle  $\theta$  with  $F$  consider another section  $F_\theta$ . Resolve the force  $P$  into  $P'$  and  $Q'$ , respectively perpendicular and parallel to  $F_\theta$ , or tension and shear on  $F_\theta$ . Then  $P' = P \cos \theta$ , and  $F_\theta$ , in area,  $= \frac{F}{\cos \theta}$ . Hence the intensity of tension stress on  $F_\theta$  is

$$p' = \frac{P'}{F_\theta} = \frac{P \cos \theta}{\frac{F}{\cos \theta}} = \frac{P \cos^2 \theta}{F} = p \cos^2 \theta.$$

As  $\cos^2 \theta$  is unity for  $\theta = \text{zero}$ , and less than unity for any other value of  $\theta$ ,  $p$ , the tension stress perpendicular to  $F$ , is the **maximum tension stress** in the test piece. Hence, if the break of the piece be due to tension, the piece will break "square off," on some normal cross-section such as  $F$ .

\* The distance by which the surfaces must separate in the tension break, to pass the range of molecular attractive forces, that is, the tensile strength forces, is about one one-hundred-millionth of an inch. It does not take long to go this distance.

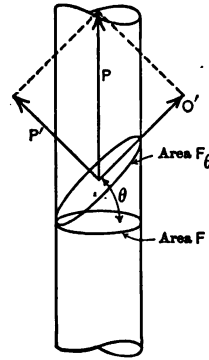


FIG. 13.

**35.** An apparent exception to the tension form of break occurs in a test piece of the form of Fig. 9B. This break is conditioned by the tension stress distribution shown in Fig. 10, and takes a concave form, hollowed into the end of the test piece, as shown in section in Fig. 14. It is easily seen that this is still a pure tension break, perpendicular to the direction of tension stress.

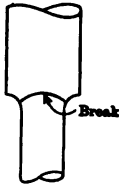


FIG. 14.

**36.** Other brittle materials in engineering use are woods, stone, cement, concrete and some of the hardened steels. The same general phenomena and the characteristic curves in all of these will be found, in tension loading, very similar to the case discussed, cast iron.

**37.** The tensile strength of stones, concretes and cements, is too low to make them useful for tensile loading in actual construction. Their tensile strengths usually lie under 1000 pounds per square inch.

Woods are fairly often used under tension (as beams, where the internal stress is partly tension) and are moderately strong. Their tensile strengths range from 5000 pounds per square inch to 30,000 pounds per square inch. Shearing along or across the grain is a more important factor in the failure of woods than the tensile strength.

Cast irons are quite commonly used in tension. The strengths run from 15,000 pounds per square inch in large castings of full gray iron, through 20,000 or 30,000 pounds per square inch in good castings of light gray iron, to about 45,000 pounds per square inch for white cast iron.

The brittle hardened steels may have any strength from zero up to 300,000 pounds per square inch depending upon the internal condition set up by the heat treatment. Cooling stresses within the material may be of the same magnitude as the intrinsic strength and so reduce the available strength to nearly nothing.

**38.** The tension test is not in general a standard or specified test for brittle materials. It is sometimes used with cast irons and is at present a standard test with cements. The wisdom of the use of the tension tests with cement will be discussed in the chapter on cements.

## CHAPTER IV

### TENSION LOADING OF DUCTILE MATERIALS

39. The ductile materials of engineering are all metals. They include wrought iron, steels, copper and the copper alloys (brasses and bronzes) and a number of aluminum alloys. In all of these ductile metals it is safe to say that at least 90 per cent of the volume is made up of crystals of the ductile type; the remainder may be of hard and brittle crystals.

40. Tension loading of the ductile materials shows at first a relation of stress and deformation purely or nearly elastic, with the magnitude of the deformations similar to that of the brittle materials, or less. **At higher loads the ductile materials have another relation of stress to deformation, a relation which is absent in brittle materials: a region where small increases of stress are accompanied by large permanent deformations.**

41. Some of the ductile materials are used as **castings**. In castings, the crystals are arranged with their long axes perpendicular to surfaces of the casting. Tension loading parallel to the surfaces (and the length of the piece) will give at first a stress-deformation relation, in this case, like that of cast iron. There is no very definite elastic limit. The loading line is slightly curved from the beginning.

42. Most of the ductile metals are **worked** materials, that is, in process of manufacture they have been forged or rolled or drawn with the result that the crystals tend to be elongated parallel to the surfaces or length of the piece, imitating somewhat the fibrous structure of wood. The worked metals usually have in the first part of their loading a considerable range of elastic action.

43. Fig. 15 shows, in *OA* and *OB*, the relative positions of the lines of first loading of worked and cast steels. *OA* for worked steel is practically a true straight elastic line. *OB* for the cast steel is not

quite straight nor quite pure elastic action. There is some permanent set even at low stresses, although these small permanent sets below the yield point are unimportant in practice.

44. The common definition of elastic action is that, during such action, stress intensity and deformation are proportional to each other.

This common definition is sufficiently good for engineering purposes. It is nearly true because of the accident that the range of elastic deformation possible in the metals is so small that it is impossible to find out whether the relation is a straight line or a curve. If we take a material, such as rubber, capable of about 700 per cent increase of length in elastic action, it will be found that the relation of real stress (actual load over actual area at each point) to deformation is a parabolic curve. The stiff-

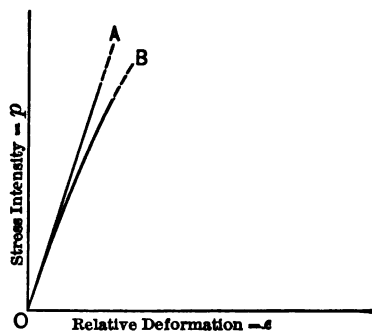


FIG. 15. Curves Showing Beginning of Loading of Rolled Steel, A, and Cast Steel, B.

ness or slope of this curve increases with the stress. The deformation range in the elastic action of the metals is only about one-half of one per cent elongation, entirely too small to detect curvature in the stress-deformation relation.

Fig. 16 shows the form of the tension stress-deformation curve for rubber. It is evident that if we confined ourselves to a very small portion of this curve near the origin, such as  $OA$ , we might consider the elastic stress-deformation relation within the limits of  $OA$  as a straight line.

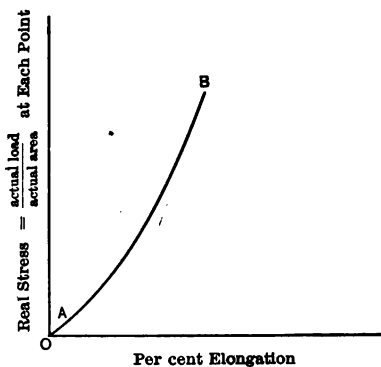


FIG. 16. Form of the Tension Stress-deformation Curve for Rubber.

45. There is a small **time factor in elastic deformations.** The bulk of the deformation occurs as instantaneously as the application of the load. There is a slight

further increase of deformation with time as the load remains, known to the physicists as the "elastic after effect."

Fig. 17 shows the nature of the "elastic after effect." A sudden loading gives the stress-deformation relation  $OA$ . Holding of the load will give a further deformation

$AB$ . Sudden unloading, followed by rest, will similarly give  $BCO$ . If the loading had been slowly applied we would have obtained the ordinary straight line stress-deformation relation  $OB$ , the dashed line. The difference of slope of  $OA$  and  $OB$  is of the order of one per cent. The "elastic after effect" is of no importance in engineering, except perhaps in the case of fatigue failure.

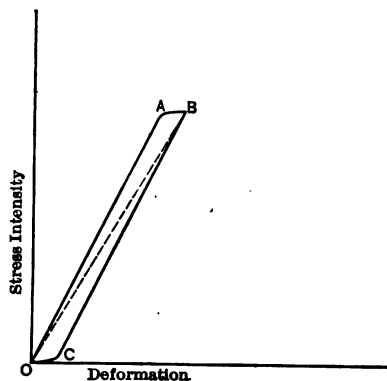


FIG. 17. To show "Elastic After Effect."

**46. The real criterion of elastic action is the perfect return of the material to the original dimensions at the end of a cycle of loading and unloading.**

**47.** When loading begins, we have throughout the metal, from crystal to crystal, the same deformation common to all. The stresses will vary from crystal to crystal, depending on the stiffness of the different crystals, the magnitude of the deformation, and the initial condition of the individual crystals. The calculated stress intensity in the material is merely an average value. The stress intensities in individual crystals are unknown, some being above and some being below the average. As the loading increases some of the most highly stressed crystals reach their elastic limit, form slip planes, and begin to take permanent deformations. When this happens we have reached the real elastic limit of the material as a whole. Unloading would leave some permanent set, though it might be very small.

**48.** Continuing the loading beyond the elastic limit of the piece, more and more of the crystals would pass their individual elastic limits. The permanent deformation of the piece as a whole would become visibly large, when nearly all of the crystals have begun to form slip planes. When this condition is first reached the load

causing it is called the **yield point load**. The **yield point** might for general purposes be defined as that stress intensity which causes a **permanent elongation of one-half of one per cent**. This definition corresponds very closely to average engineering practice. Any general definition of yield point must necessarily be arbitrary.

49. We may say approximately that up to the yield point we are dealing mostly with the **elastic deformation** of the piece caused directly by the **tension stresses**. Beyond the yield point we are dealing mostly with **permanent deformation** of the piece caused directly by **shears which are secondary to the tension**. It is necessary to know the relative directions and magnitudes of the tension and shear stresses.

50. Internal stresses of tension or compression which have their maximum intensities in the direction of the external tension or compression loads, or in the plane of a bending moment, we will call **primary** tension or compression stresses. Internal shear stresses which have their maximum intensities in the direction of an external shear force or in the plane of a torsion moment we will call **primary** shears. When the maximum intensity of a stress lies at some angle not zero or 180 degrees with the external load causing it, that stress will be called a **secondary** stress. Primary stresses are the obvious stresses set up by their loadings. Secondary stresses are not obvious at casual inspection of a loading. They are revealed when we place an analysis plane in any direction through the material and examine the stresses on it.

Primary and secondary stresses are **equivalent systems** of internal stresses, not a **simultaneous double system**. In any loading it is necessary to locate the directions and magnitudes of both **maximum tension (or compression) stresses** and **maximum shear stresses**, because failure of the material may occur by either shear or tension, or both in sequence.

51. In Fig. 18, showing a portion of a tension piece, we would usually calculate the tension stress intensity by dividing the load  $P$  by the area  $F$  of the cross-section  $AB$ , perpendicular to the load. This computes the intensity of the primary tension stress acting in the piece. Instead of  $AB$  consider any plane such as  $AC$  making the angle  $\theta$  with  $AB$ . On the plane  $AC$  resolve the load,  $P$ , into the components



$P'$  perpendicular to  $AC$  and  $Q'$  parallel to  $AC$ .  $P'$  is a tension on  $AC$  and  $Q'$  a shear on  $AC$ . The stress intensities on  $AC$  due to these forces are found by dividing force by area. The area of  $AB$  being  $F$ , the area of  $AC$  is  $F'$  equal to  $\frac{F}{\cos \theta}$ . In the parallelogram of forces  $P'$  equals  $P \cos \theta$ , and  $Q'$  equals  $P \sin \theta$ . Then we have:

Maximum tension stress on  $AB$  equals  $p = \frac{P}{F}$ ; tension stress intensity on  $AC$  equals

$$p' = \frac{P'}{F'} = \frac{P \cos \theta}{\frac{F}{\cos \theta}} = \frac{P \cos^2 \theta}{F} = p \cos^2 \theta;$$

and shear stress intensity on  $AC$  equals

$$q' = \frac{Q'}{F'} = \frac{P \sin \theta}{\frac{F}{\cos \theta}} = \frac{P \sin \theta \cos \theta}{F} = p \sin \theta \cos \theta.$$

We may deduce, since  $\cos^2 \theta$  has its maximum value of unity when  $\theta$  is zero, that the tension stress,  $p$ , on  $AB$ , is the maximum possible tension stress, and therefore is the primary stress. The tension stress in any other direction than perpendicular to  $AB$  may be neglected, if the material is isotropic,\* because any critical action from tension stressing must happen where the tension stress intensity is maximum, that is, on  $AB$ .

To find the dangerous place for shears make the trigonometric substitution of  $(\frac{1}{2} \sin 2 \theta)$  for  $(\sin \theta \cos \theta)$  in the equation above. Then  $q' = \frac{1}{2} p \sin 2 \theta$ .  $\sin 2 \theta$  has a maximum value of unity when  $2 \theta$  is 90 degrees, or  $\theta$  is 45 degrees.

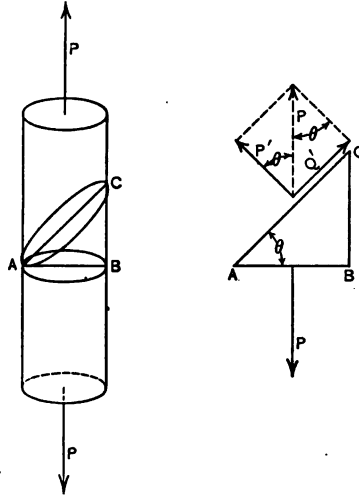


FIG. 18.

Hence at 45° to the maximum

\* With the exception of wood, engineering materials are nearly isotropic, and may ordinarily be so considered.

tension stress, on planes such as  $AC$  at  $45^\circ$  to the length of the piece, we have a secondary (maximum) shear stress  $q'$  numerically equal to  $\frac{1}{2} p$ .

Up to the yield point we are dealing with the actions of the tension stress. Beyond the yield point, or more strictly, whenever there is a permanent deformation, we are dealing with the action of the secondary shear stress.

52. In the soft steels the bulk of the structure (90 per cent or more of the volume) is of ferrite crystals. These crystals are isotropic, having the same stiffness in all directions through their structure. Hence the orientation of the crystal structure with regard to the dimensions of the piece as a whole has no effect upon the stress intensities set up in the crystals by the common deformation. As a result the stress intensities from crystal to crystal in a steel are of very nearly the same magnitude. The only outstanding reason for differences of stresses from crystal to crystal, or part to part, of the piece, is the initial condition of internal stress due to previous history. The piece may have had irregular cooling or uneven working or other such causes of internal stress. It is usually the case in steels that the great proportion of the ferrite crystals reach their elastic limits at about the same time. This causes the yield-point load to be marked by a sudden considerable elongation of the whole piece at practically constant load.

53. Fig. 19 shows the course of the tension stress-deformation curve for a ductile steel. The first permanent set or elastic limit

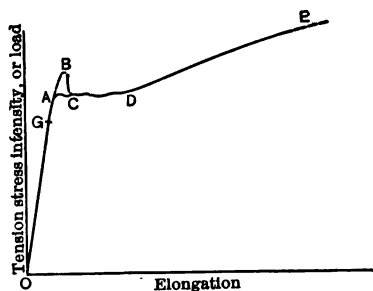


FIG. 19. Tension Stress-deformation Curve of Soft Steel, through the Yield Point.

would be detected at  $G$ . The yield-point load is practically constant on  $ACD$ . The location of the elastic limit is not important in engineering practice. The important thing is the yield-point load. The placing of the elastic limit is exceedingly variable in testing. It depends partly on the accuracy of the measurements and partly on the perfection of setting up of the test piece,

so as to secure concentric loading. The more accurate the measure-

ments the lower will we place the point *G* where the first permanent elongation can be detected. *G* is usually placed at the first perceptible departure of the curve from straight line action. Eccentric loading causes non-uniform stress through the material, bending the piece so that one side reaches the yield point ahead of the other. The main effect of this is to lower the placing of the elastic limit.

54. There is a considerable **time effect in the permanent yielding** of the piece. With machinery sensitive and accurate to rapid changes, it will be found that the load momentarily runs to some peak such as *B* (Fig. 19) and then drops sharply to the irregular level *CD*. The ordinary testing machine is too slow, has too much inertia in its weighing system, to catch this peak *ABC*. It reports instead *ACD*, and follows only the grosser changes of load as the elongation increases. The elongation of the piece, while going through the change *ACD*, is not uniform over the entire length. It consists of a successive series of local elongations. First one spot, then another, on account either of local smallness of section or relative softness of material, reaches its yield-point stress and begins to stretch. Each part in turn comes to the same action which characterizes the piece as a whole. After a certain amount of deformation at nearly constant load, a condition is reached where continuation of the deformation requires increasing load (more strictly, increasing stress intensity). The branch *DE* of the curve shows this for the whole piece of steel. The series of successive local yieldings shows in the stress-deformation curve in the wavy portion *ACD*.

55. The ductile metals other than steels usually have either non-isotropic crystals or very large initial internal stresses. These conditions prevent the fairly simultaneous yielding of all parts of the structure. The beginnings of yielding of the individual crystals are distributed over a considerable range of the loading. So in brasses and bronzes, for instance, we miss the decided yield-point action of the steels.

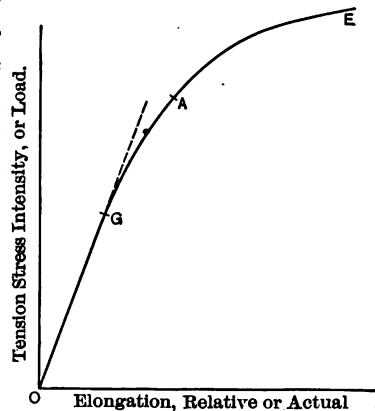


FIG. 20. Tension Stress-deformation Curve of Non-ferrous Ductile Metals.

Fig. 20 shows the type of curve found. If the metal has been drawn or rolled there is probably an elastic range  $OG$ . Beyond  $G$ , the curve changes gradually, through the yield point  $A$ , to  $E$ , which corresponds to the point  $E$  (Fig. 19) for steels. The yield point  $A$  is not marked by a period of constant load as with the steels, **but has to be arbitrarily placed** at some elongation, preferably one-half of one per cent.

56. For commercial determinations, the yield point of steels may be easily found in most cases by what is known as the "drop of beam" method. In the ordinary screw type of testing machine, the test piece is stretched at a constant time rate of elongation. The attendant moves the weighing poise on the scale beam of the machine so as to keep the beam in balance. Along the portion  $OGA$  of the curve, (Fig. 19), the load rises rapidly at a constant rate. The attendant correspondingly gets the poise on the weighing beam moving outward at a constant rate. When the yield point is reached, there is a period where continued elongation of the piece occurs at constant load. The result is that the attendant overruns slightly the poise position for the yield load  $ACD$ , making the beam drop momentarily. The attendant then notes the beam reading of this load as the yield point, without interrupting the progress of the test. It is apparent that this "drop of beam" method can be applied only in cases where there is a sharp change of direction in the stress-deformation curve at the yield-point load.

57. Eccentric loading does not greatly affect the determination of the yield-point load. At or beyond the yield point, the rate of change of load against deformation is small enough so that the minor stress differences between the sides of a piece, which is slightly bent, disappear. The bending may be either initial lack of straightness or from eccentric loading. Also, the initial yielding of a bent or eccentrically loaded piece is on the inside of the bend in such a direction as to remove the eccentricity. The piece straightens out and automatically takes its loading concentrically as soon as there has been any appreciable yielding. It follows that initial bending or eccentric loading of the piece have little effect on the yield-point determination, and still less effect on the correctness of parts of the stress-deformation curve beyond the yield point.

58. Rate of loading comes into the determination of all loads

beyond the elastic limit. The time factor in permanent deformation is the cause. The more rapid the loading, the higher is the stress intensity for arriving at a given permanent deformation. The stress intensity is not, however, proportional to the rate of loading. This time effect is relatively a minor matter. The "faking" of strengths by increasing the rate of deformation in testing cannot be very great, although it is large enough to be guarded against by specifications.

59. While the material is going through the yield-point region, and slightly beyond it, the main action within the crystals is the formation of increasing numbers of parallel slip planes. The slip planes first formed lie on, or very close to, the planes of maximum shear stress intensity in the material.

Continued deformation beyond the yield point brings in two actions, both requiring increasing stress with increasing deformation. First, there is increasing slip along the initially formed slip planes. After the beginning of motion this increasing slip is opposed by an increasing friction. At its beginning only can the slip on a slip plane be considered substantially frictionless. Second, new slip planes are forming at angles further and further removed from the planes of maximum intensity of shear stress. The general effect is a continued increase of real stress intensity with continued elongation until break of the material occurs in some manner.

60. Up to the yield point, the decrease in diameter or thickness of a piece of material in tension loading is unimportant: so small in fact that it is rather hard to measure with accuracy. In elastic loading the relative increase in length of the piece is greater than the relative decrease in cross-section area, so that the volume of the material **very slightly** increases. Permanent deformations involve little or no volume change. They consist merely of internal displacements of the material. The comparatively large elongations at and after the yield point are accompanied by appreciable decreases of diameter or cross-section area of the material. For some distance past the yield point, the change in cross-section is fairly uniform along the length of the piece. Sooner or later, the decrease of diameter or cross-section area becomes localized at some one spot. This action is called "necking." When this occurs there is a brief period of rapid change at that one place and then the piece breaks at that place.

61. In engineering practice we put into service a piece of material of known initial section. We want to know what actual load this piece will carry. For engineering purposes, therefore, stress intensities in tension loading are calculated by dividing the actual loads by the initial cross-section area. This disregards the change of cross-section which occurs at and after the yield point. The engineering stress intensity, actual load divided by initial area, is after the yield point a purely nominal quantity. The real stress intensity, which we will hereafter call the physical stress intensity, must take account of the change of cross-section and must be found by dividing each actual load by the corresponding actual cross-section area. Whether we calculate the engineering or physical value of the stress intensity is decided by the application we intend to make of the result.

62. Up to the point in the loading where the deformation localizes, the elongation is uniformly distributed over the length of the piece. If we measure the total elongation and divide it by the initial length we obtain a value of the relative elongation,  $\epsilon$ , which is not only the average value for the piece, but also the true value for each portion of the piece. After the deformation localizes or, technically speaking, the piece begins to "neck," the average elongation,  $\epsilon$ , is no longer a true value for any part of the piece. It is too large for parts outside of the "necked" portion and too small for parts in the necked portion. To get at the true value of relative elongation in the center of the necked section, it is necessary to proceed indirectly by means of the change of cross-section areas. Let  $F_o$  be the initial cross-section area and  $F_m$  be the minimum cross-section area at any stage of the loading. Using the fact that permanent deformation causes practically no volume change, we have  $F_o \times L_o = F_m \times L = \text{volume of piece}$ . Hence the length,  $L$ , which the piece would have taken if it had remained cylindrical and everywhere reduced to the cross-section area  $F_m$ , is  $L_o \times \frac{F_o}{F_m}$ . The increase in length then would have been  $(L - L_o)$ , or  $L_o \left( \frac{F_o}{F_m} - 1 \right)$ , and the relative elongation would have been the physical

$$\text{value of } \epsilon = \frac{L_o \left( \frac{F_o}{F_m} - 1 \right)}{L_o} = \left( \frac{F_o}{F_m} - 1 \right).$$

63. We have shown now that the stress calculations after the yield point may be made in two different ways: actual load divided by initial area for the "engineering" value, or actual load divided by actual area for the "physical" value. Similarly, the deformation calculations may be made in two ways: the actual total elongation divided by initial length equals average or "engineering" value of relative elongation, and  $\left(\frac{F_o}{F_m} - 1\right)$ , ratio of initial area to actual minimum area, minus 1, gives the "physical" relative elongation which corresponds to the physical stress value.

64. The curves of stress vs. deformation by the two methods of calculation are shown in Fig. 21. *OAEFK* is the engineering curve. *OAXY* is the physical curve. The physical curve lies above the engineering curve because the actual areas of cross-section are less

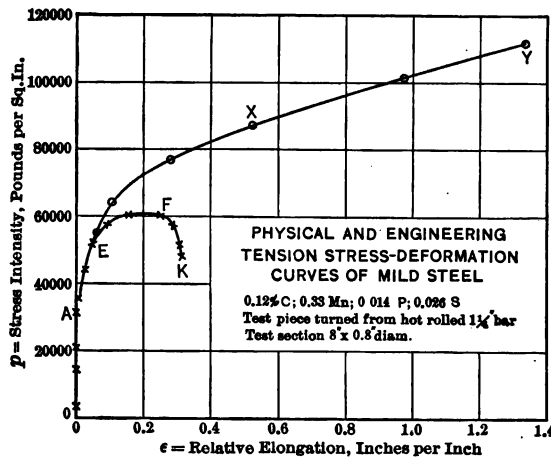


FIG. 21.

than the original. Toward the end of the loading the local concentration of deformation at the necked section of the piece makes the physical value of relative elongation considerably greater than the engineering value. Up to the yield point, *A*, the two curves coincide because the change of cross-section area from the original is negligible.

65. When finding properties of the material for use in engineering design or specifications we calculate values by the method above called "engineering." For study of the properties of the material

itself, as in comparing properties in tension with properties in torsion or transverse loading, we must use the "physical" method of calculation. Again, when discussing the action of the material in rolling or wire drawing, pressing and similar processes, we must use the physical not the engineering method as the basis of our calculations. The engineering method of calculation has one application only,— to figure back from the loads carried by a test piece of known initial cross-section, to the loads which would be carried by a tension member of the same material in an actual structure at similar points in its failure, if it were loaded to failure.

66. The successive stages in the deformation of a tension piece of a fully ductile metal are shown in Fig. 22. 1-1 shows the initial cylindrical section. Just after passing the yield point the piece would appear as 2-2, slightly less in diameter and greater in length. Going beyond the yield point for a considerable period the piece



FIG. 22.

continues to stretch uniformly to the appearance of 3-3. The total load on the piece is the product of the actual stress intensity by the actual cross-section area. The actual or physical stress intensity is continually rising with the increase of elongation. The area of the cross-section is continually decreasing. There usually comes a point where the decrease of the area becomes a more important factor in the size of the product (actual stress intensity times actual cross-section area) than the increase of stress intensity. That is, the total load on the piece reaches a maximum value, and subsequently decreases. This stage of the loading gives the point *F*, the "maximum load point" on the engineering curve, Fig. 21.

At or just after the maximum load the piece begins to neck. The elongation and reduction of area become localized. After this begins, the parts of the piece outside of the necked portion undergo no further change. If 3-3, Fig. 22, represents the condition of the piece at the maximum load point, the condition just before break occurs at *B* is shown by the dotted outline 4-4. 4-4 coincides with 3-3 except for the changes due to the necking at *B*.



This stage of the test shows on the engineering curve as the dropping end  $FK$ , Fig. 21. The form of  $FK$  is exceedingly variable, depending on the ratio of length to diameter of the original test section. When we change our calculations to the physical basis,  $FK$  of the engineering curve becomes  $XY$  of the physical curve. This is characteristic of the material and is independent of the form of the test piece. The actual stress intensity carried by the material increases always up to the break point. The drop of stress intensity from  $F$  to  $K$  on the engineering curve is due to the method of calculation. The "engineering" stress at this stage of the loading is a purely nominal quantity.

67. The tension test terminates with one or the other of two forms of break. The piece may break squarely across like the brittle materials. This is shown in Fig. 23A. The usual forms of break are Fig. 23B and C. Fig. 23D is occasionally found. These latter are called "cup and cone" breaks. They start at the outside of the piece by shearing the material at about 45 degrees to the axis. This shear break works gradually inward, rapidly reducing the area of metal

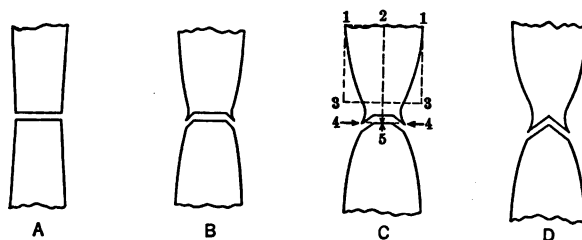


FIG. 23. Sections through Break Forms of Ductile Metals in Tension.

remaining to carry the load of the piece. This results in a rapid rise of actual stress intensity in the inner portions remaining and ends in tension breaking of these inner parts. The tension break is perpendicular to the tension stress and forms the flat part of the cup and cone as the shear break forms the slanting sides. The break starts at the outside because the stress intensity is higher there than inside. In Fig. 23C, 1-1-3-3 represents the original outline of one-half of the "neck" of the piece. The sides 1-3 are elongated to 1-4, while the center line, originally of the same length as 1-3, elongates only to 2-5. 1-4, measured on a curved line, is appreciably longer than 2-5. Hence,

in the necked portion of the piece, deformation of the material is a variable through the cross-section, greatest in the outer skin. Correspondingly the stress intensity must be variable through the cross-section, greatest in the outer skin. Therefore the break must start at the outside and work inward. We know that with a given intensity of tension stress there goes a secondary shear stress, at 45 degrees with the tension, and numerically of one-half the value. If the breaking strength of the material in pure tension stress is less than twice the breaking strength in shear stress, the break will start and finish by tension as in Fig. 23A. Usually in this case there is little or no necking, and the material should be classed as semi-ductile rather than ductile. If the breaking strength of the material in pure tension exceeds twice the breaking strength in shear, then the break begins by shearing in the outer surfaces, giving the cup and cone form. Just how far the shear break goes inward, before the tension break finishes the piece, depends on the excess of the tension strength (against pure tension break) over twice the shear breaking strength of the material. The shearing inward is progressive, taking measurable time. A pure tension break is practically instantaneous. Hence, when the stress intensity on the remaining core of material reaches the breaking strength in pure tension, the piece finishes instantly with tension break, instead of shearing deeper.

68. Because the stress intensity across the necked section is not uniform and the variation from uniformity is unknown, the calculation of actual load divided by actual area at break does not give the real stress intensity which starts the break. This means that the part XY of the physical curve, Fig. 21, is somewhat untrue in its stress values. The calculations are true until necking begins. The error increases as the necking becomes more decided. The flatness of the curve shows that it takes a considerable difference of deformation to make a small difference in the stress intensity. Hence the error in the calculations cannot be very large.

69. Because the real stress intensity in the outer surface of the necked portion at the beginning of break is not exactly known and because also the relation of tension stress intensity to shear stress intensity is modified by friction of slippage in the shear break, the calculation of the breaking strength in shear from the breaking

strength in the tension test is possible only in an approximate fashion.

70. The engineering measurements of tensile properties include yield-point strength, and maximum-load strength, obtained by dividing the loads at each place by the original area. The breaking load should be noted, but often is not. English engineers often speak of the "ultimate strength" of the metal in the sense in which the American engineers speak of the "maximum strength." "Ultimate strength" in technical usage does not mean the breaking strength. In ordinary testing no record is kept of the changes of cross-section area during the progress of the test. The calculation of the physical curve is then impossible.

In addition to knowing how strong the material is we want to know how much it changes in shape both under elastic loading and when loaded to break. The changes in shape in elastic loading determine the stiffness or modulus of elasticity of the material. Knowledge of this is occasionally necessary in design for the computation of the rigidity of a structure.

The deformation of the material at break may be measured in two ways: by the change in length, or by the change in cross-section area. On account of the necking of the piece these two methods are not equivalent. The ratio of total elongation at break to original length of piece, expressed in per cent, is called the "ductility" of the metal. The ratio of the difference between the initial and final cross-section areas to the original area is called the "reduction of area." It is likewise given in per cent. The ductility is much more generally used in specifications than is the reduction of area. This practice is unfortunate because the ductility number for a given material is very much a function of the shape of the test piece used, while the reduction of area measurement is practically independent of the shape of test piece.

71. Ductility is determined in commercial testing by measuring the initial length of the test piece and after the break putting the broken ends together and measuring the final length. It would require the running of a large number of test pieces with various ratios of initial length to original diameter, to show with this method of measurement the effect of shape of test piece upon ductility. A much sim-



$\left( \frac{\text{original area}}{\text{final area}} - 1 \right)$ . As the original length of test piece is increased, with the same diameter, the ductility figure rapidly decreases. The ductility curve becomes asymptotic to the percentage stretch which characterizes the parts of the broken test piece remote from the necked section.\*

It has been found experimentally that test pieces of the same material and having the same ratio of length to diameter of initial test

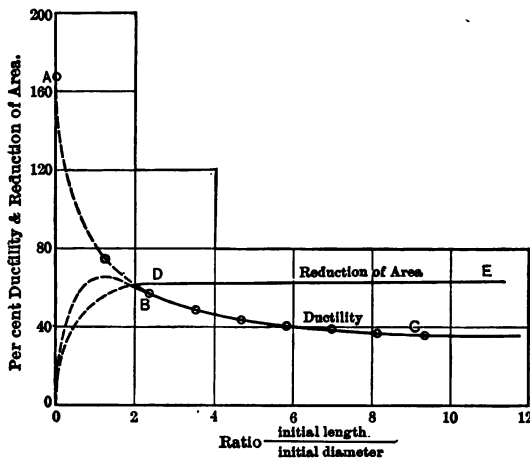


FIG. 25. Ductility vs. Shape of Test Piece for Test Piece of Fig. 24.

section will give the same ductilities, although the absolute size of the pieces may vary through wide limits. For example, test pieces of a certain kind of steel two inches long and one-half inch in diameter (a four to one ratio of length to diameter) give 30 per cent ductility.

\* An approximate equation for the curve of Fig. 25 can be made by assuming that the elongation consists of two parts. One part, outside the necked portion, is proportional to the initial length of the test piece. The other part, in the necked section, is proportional to the initial diameter of the piece. In algebraic statement,

$$\Delta L = aL + bD. \text{ Hence } \frac{\Delta L}{L} = \text{ductility} = \frac{aL + bD}{L} = a + b \frac{D}{L}$$

$a$  and  $b$  are constants for the material;  $L$  is length of test piece;  $D$  is its diameter. By marking off on a test piece two initial lengths, data will be found from that test for determination of  $a$  and  $b$ .

So, also, will pieces of that same steel, four inches long by one inch in diameter or eight inches long by two inches in diameter give 30 per cent ductility. The ductility figure depends on two things: first, the material itself; second, the shape of the initial test section as measured by the ratio of length to diameter or, more generally, the ratio of initial length to the square root of initial cross-section area. A ductility figure is meaningless unless the dimensions of the test section used are simultaneously given. Likewise a ductility specification is not binding without simultaneous specification of the dimensions of the test piece to be used.

72. The device of finding from one single test piece the ductilities that would result from a series of pieces of different initial lengths, fails when the ratio of original test length to original diameter is less than about three. Let Fig. 26 represent a series of test pieces in which the test length is made successively shorter until in *E* the test

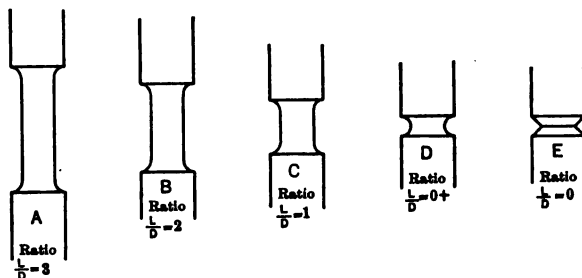


FIG. 26.

length becomes zero. The ductility measurements from this series would not plot on the curve *CBA*, Fig. 25, but on the curve *CBO*. When the test length becomes very short compared with the diameter, the material in the ends of the test piece furnishes to the material in the test section sidewise support against reduction of area. In the extreme case, ductility vanishes as the test length reaches zero. If on the same series of pieces we measure the reduction of area we will get the curve *ODE* likewise going to zero when the test length becomes zero. In this case, as soon as the test length becomes great enough so that the center of the test section is free from constraint by the ends of the test piece, the reduction of area measurement becomes constant,

and is independent of any further increase of the test length. Comparison of the curves of reduction of area and ductility as functions of shape of test piece, in Fig. 25, will show immediately the advantage of reduction of area rather than ductility for the engineering measurement of the deformation of the material under test.

73. The engineering stress-deformation curve of a material also changes with shape of test piece. While the test length exceeds three times the diameter, the change of test length affects only the tail of the curve beyond the maximum point  $F$ , Fig. 27. The longer the test section the shorter the tail  $FK$ . When the test section is so shortened that the material in the test section is constrained against deformation by support from the ends of the piece, the whole curve in Fig. 27 is raised and shortened, giving such curves as  $OAW$ ,  $OAX$ ,  $OAY$  and  $OAZ$ , Fig. 27.  $OAZ$  corresponds approximately to Fig. 26D or E as shape of test piece.

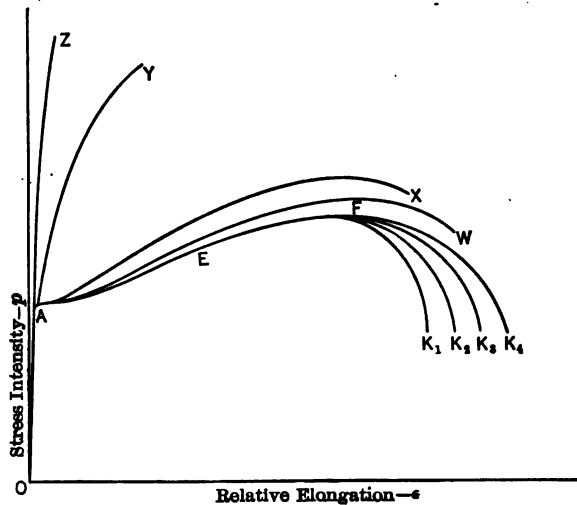


FIG. 27.

74. Because the ductility measurement and, to a lesser extent, the reduction of area and strength measurements are functions of the form of the test piece, as well as of the material itself, it is necessary to specify and to standardize the forms and dimensions of test pieces to be used. For test pieces of round section there are

two standards. In most common commercial use is the piece with test section 2 inches long by 0.5 inch diameter, or a ratio of 4 to 1, length to diameter. The International Association for Testing Materials has recommended the less common form with 8-inch test length and 0.8 inch diameter, a ratio of 10 to 1. Unless some special reason prevents, one or the other of these standards should be used in all tension testing. The only difference in results between them is the ductility measurement, which is considerably larger for the smaller piece.

For rectangular cross-sections which occur in test pieces representing sheet or plate steel or other metals, definite standards of dimensions have now been specified by the American Society for Testing Materials, and other authorities. Unless these standards are adhered to, ductility and reduction of area figures and, to a lesser extent, strength figures are valueless.\*

75. The tension test is the standard and practically universal test for quoting and specifying the properties of the ductile metals. The unqualified statement that the strength of a metal is a certain number of pounds per square inch is understood to mean the "engineering" maximum tension strength. The tension test is so universally satisfactory because the measurements of yield-point strength, maximum strength and breaking strength, and the deformation of a material in the tension test really carry, concealed within them, information as to the shear strength and deformations which the material will take. The tension test therefore enables us, directly or indirectly, to foretell the action of ductile material not only in tension loading but also in compression, transverse and torsion loading. The subject of the numerical cross-relationships of the properties of the materials under the different systems of loading will be developed in more detail in a subsequent chapter.

\* Standard specifications for testing any materials, so far as there are standards, can be obtained from the American Society for Testing of Materials.



## CHAPTER V

### TORSION LOADING

**76.** A torsion loading is any loading which tends to twist a piece. It consists, therefore, of forces or components of forces which are directed perpendicular to the length of a piece, and which do not pass through the center of the cross-section of the piece. The perpendicular distance from the axis of the piece to the line of action of the force is the **lever arm** of the force. The product of the length of the lever arm and the magnitude of the component of the force in a plane perpendicular to the axis of the piece, is the **twisting moment**,  $M$ . (Force components in planes parallel to the length of the piece cause tensions, compressions or bending moments.)

**It is the magnitude of this product (force  $\times$  lever arm) which measures the torsion load.** The factors may be distributed in any way. Large force and small lever arm have the same effects as small force and large lever arm if the products (force  $\times$  lever arm) are the same.

**77.** Torsion loading occurs in practice in shafting. Pure torsion loading is rare. There is usually complication by transverse loading, as in crankshafts, or line shafting in mills. Usually the torsion moment is constant along considerable lengths of the piece. A torsion moment varying along the length of the piece occurs within the portions where load is being applied or removed, as where a shaft passes through the bore of a pulley or at the crank pins of a multi-cylinder engine. Design is made for those portions of the shaft carrying constant moment. Hence in testing and in the discussion to follow we deal almost entirely with the case of constant torsion moment.

**78.** Let Fig. 28 represent a portion of a solid round shaft loaded with a torsion moment,  $M = P_1 \times a_1 = P_2 \times a_2$ . To have equilibrium of moments against rotation we must have equal and opposite moments at the two ends of the piece. To secure complete force equilibrium on the piece there must be two bearing forces  $P_1'$  and  $P_2'$  equal and opposite respectively to the forces  $P_1$  and  $P_2$ . Through

any point  $O$  of the axis, between the two loads, pass a perpendicular plane  $ABCD$ . Then it is evident that the action of the twisting moments tends to slip the part at one side of  $ABCD$  past the part

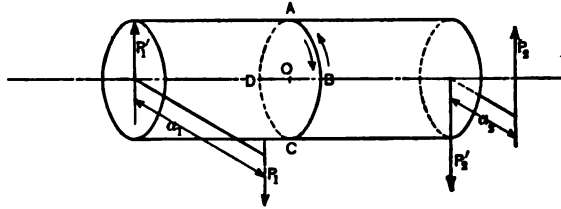


FIG. 28.

on the other side. Each part tends to rotate in the direction indicated by the arrows at  $AB$ . Therefore the internal stresses set up by the torsion consist, in part at least, of shear stresses in planes perpendicular to the axis of the piece. There are no external loads directed parallel to the length of the piece. Hence there are no tension or compression stresses perpendicular to such cross-section planes as  $ABCD$ .

79. For further study of the internal stresses, isolate a small cube of material by passing a pair of perpendicular planes such as  $DB$  intersected by planes through the axis, as indicated by Fig. 29. A cube so obtained is shown, magnified, in Fig. 30. From the material to the left of it, we know that the cube receives a shear load  $Q_1$  on the

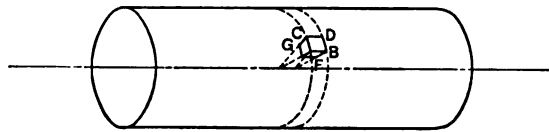


FIG. 29.

face  $ACGE$ . From the material to the right of it we know that the cube receives a shear load  $Q_2$  on the face  $BDHF$ . These two loads are indicated by the previous discussion of the slip actions on the plane  $ABCD$ , Fig. 28. If the cube is in equilibrium against sidewise motion  $Q_1$  and  $Q_2$  must be equal.  $Q_1$  and  $Q_2$  constitute a couple tending to rotate the cube. To maintain equilibrium against this rotation, there must be two other shears  $Q_3$  and  $Q_4$  acting on the cube on the faces  $ABFE$  and  $CDHG$ .  $Q_3$  and  $Q_4$  must be equal to each other and

also equal to  $Q_1$  and  $Q_2$ . The shears on such an elementary cube, therefore, occur not only in pairs constituting couples, but in **pairs of pairs**. There must be a belt of four shears of equal magnitude around the cube so disposed as to balance against both rotation and translation of the cube. These shears represent the force actions of adjacent material upon the faces of the cube. If at any point in a material there is a shear stress intensity of  $q$  pounds per square inch in a given direction, then there is some plane at  $90^\circ$  to this direction which likewise has the same intensity of shear stress. An illustration is in the equal forces  $Q_1$  and  $Q_3$  on the equal faces  $ACGE$  and  $ABFE$  of the cube in Fig. 30.

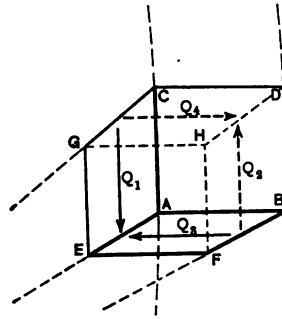


FIG. 30.

80. From this study of the elementary cube, we see that the shear stresses set up by torsion are directed in planes perpendicular to the length of the piece and also in planes parallel to the length of the piece. At any given point in the material the intensities of the crosswise shear stresses and the lengthwise shear stresses are identical.

81. The external deformation of a piece in torsion loading is a twisting of one end with regard to the other, or of any section with regard to other sections along the piece. In Fig. 31, let the end C

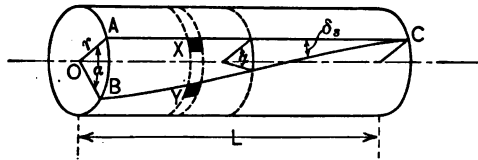


FIG. 31.

of the piece be fixed while the other end is twisted. Let a mark  $CA$  be drawn on the surface of the piece, parallel to the axis, before loading begins. After twisting the left-hand end of the piece through the angle  $AOB$ , equal to  $\alpha$ , the line originally  $CA$  has become  $CB$ . Assuming that the piece is cylindrical in shape and of uniform material,  $CB$  will be a helix line on the surface of the piece, uniformly inclined to  $CA$  at the angle  $\delta_s$ . The angle  $\alpha$  varies along the piece in direct proportion to the distance from  $C$ .

The external measurements for torsion loading consists of  $M = P \times a$  for the load measurement, with corresponding values of  $\alpha$ , the angle of twist. The diameter of the piece and the length along the piece on which  $\alpha$  is measured must be found at the beginning of the test.

82. At X, Fig. 31, is shown the initial condition of the face of an elementary cube of the material. Its condition after twisting is shown

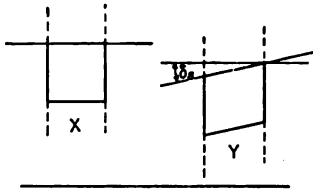


FIG. 32.

at Y. Picked out and magnified, these are shown in Fig. 32. Then it is seen that  $\delta_s$  is the angle of unit deformation of the material. (Compare Fig. 2 and its discussion in Section 6.)  $\tan \delta_s$  is the unit deformation in the surface of the piece caused by the shear stress

intensity acting in the surface.  $\tan \delta_s$  can be found directly from  $r$ ,  $L$  and  $\alpha$ , the dimensions of test section and the angle of twist, Fig. 31.

83. The relations are shown in Fig. 33.  $AOB$  is taken from the end of the piece, Fig. 31, and  $ABC$  is the unrolled or developed surface of the cylinder. On development  $ABC$  becomes a right-angled triangle with the helix line  $BC$  as hypotenuse. The arc  $AB$  has the length  $r \times \alpha$ , when  $\alpha$  is expressed in radians. In the triangle  $ABC$ ,  $AB$  has the length  $L \cdot \tan \delta_s$ . Hence  $L \cdot \tan \delta_s = r \cdot \alpha$ , or  $\tan \delta_s = \frac{r}{L} \cdot \alpha$ .

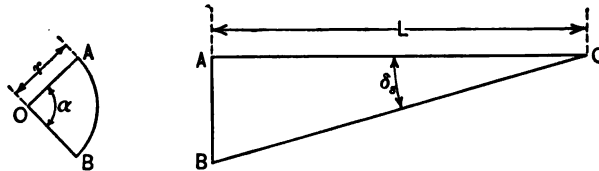


FIG. 33.

84. With this same method of calculation we can do more than find the deformation in the surface of the piece. We can find out what the deformation is anywhere within the piece. At any given stage of the loading,  $\alpha$  has a definite value. With  $\alpha$  and  $L$  fixed, use

a variable radius  $\rho$  in place of  $r$  in the equation  $\tan \delta_s = \frac{r}{L} \cdot \alpha$ . Then

we have the general equation  $\tan \delta = \frac{\alpha}{L} \cdot \rho$ . This equation shows

that the unit deformation by the shear stresses is not constant through the piece, but varies in direct proportion to the radius from the center to the surface.

85. In this variation of deformation through the cross-section of the piece, we have a condition entirely unlike that of the tension loadings, where the deformation is generally of uniform value throughout the piece. The equation  $\tan \delta = \frac{\alpha}{L} \cdot \rho$  is the key to mathematical discussion of the torsion loading. It is derived from purely geometrical considerations and is true at all stages of the loading from the beginning to the break point, regardless of the magnitude of the deformations, whether they are elastic or permanent.

86. If the loading is elastic, the shear stress intensity  $q$  is proportional to  $\tan \delta$ . The key equation  $\tan \delta = \frac{\alpha}{L} \rho$  shows that the deformation  $\tan \delta$  is proportional to the radius. Therefore in elastic loading stress intensity  $q$  is proportional to the radius, varying from zero at the axis of the piece to a maximum in the outer surface. The stress intensity  $q$  at any radius  $\rho$  is in proportion to the stress intensity  $q_s$  in the outer surface as  $\rho$  is to  $r$ .

87. As in operation the working stresses in the material are always well below the elastic limit, we deal in actual structures with elastic loading and therefore base our formulas on the condition that stress intensity is proportional to radius. We must never forget that the formulas derived on this basis are no longer true when the elastic limit is passed.

88. Let Fig. 34 show a cross-section through the piece. Consider a ring of material of radius  $\rho$  and width  $d\rho$ . This ring is built up as illustrated of small parts, one of which is represented separately, blackened. Each small part has acting upon it a shear stress  $dQ$  in the plane of the cross-section and perpendicular to the radius out to the part. The radius is the lever arm of the force on the part. All parts of the ring in Fig. 34, having the same radius, must have the same value of  $\tan \delta$  and the same value of  $q$ . The area of the ring is equal to circumference times width,

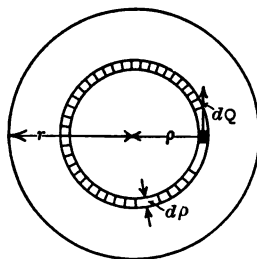


FIG. 34.

$2 \pi \rho \cdot d\rho$ . The total shear force on the ring is equal to stress intensity times area and is equal to  $q \cdot 2 \pi \rho \cdot d\rho$ . The moment arm of this shear force is  $\rho$ . Hence the element of moment from the shear stress on the ring is  $\rho \cdot q \cdot 2 \pi \rho \cdot d\rho = dM$ . For elastic loading,  $\frac{q}{q_s} = \frac{\rho}{r}$  or  $q = q_s \cdot \frac{\rho}{r}$ . The value of  $q_s$  is the quantity desired. Substituting the value of  $q$  in terms of  $q_s$ ,  $dM = \rho \frac{q_s}{r} \cdot 2 \pi \rho \cdot d\rho = \frac{q_s}{r} \cdot 2 \pi \rho^3 \cdot d\rho$ . Integrating between the limits zero and  $r$  for  $\rho$  (solid circular shaft),

$$\begin{aligned} M &= \int dM = \frac{2 \pi q_s}{r} \int_0^r \rho^3 d\rho \\ &= \frac{2 \pi q_s}{r} \left[ \frac{\rho^4}{4} \right]_0^r = \frac{2 \pi q_s}{r} \cdot \frac{r^4}{4} \\ &= \frac{\pi r^3}{2} q_s. \end{aligned}$$

$$\text{Solving for } q_s, \quad q_s = \frac{2}{\pi r^3} \cdot M = \frac{16}{\pi d^3} \cdot M = \frac{5.093}{d^3} \cdot M.$$

89. These are the ordinary formulas for the usual round, solid shaft. It is to be noted that the **strength of the shaft**, measured by  $M$ , **varies as the cube of the radius or diameter** when  $q_s$  is held constant. This is to be contrasted with the case in **tension loading** where the **strength of the piece varies with the square of the diameter or with the area of the cross-section of the piece**. In designing a series of structures of the same kind, but of different sizes, containing both tension and torsion members, one could not go from one size to another by using a proportional alteration in all dimensions, because of this different effect of diameter on strength in tension and torsion members. Each size must be checked from its own load or power requirements.

90. If a shaft is hollow the formula for its strength is readily obtained by suitable change of the limits of  $\rho$  in the integration above. Instead of zero and  $r$  use  $r_2$  and  $r_1$  for inner and outer radii of the hollow piece. This gives

$$M = \frac{2 \pi q_s}{r_1} \left[ \frac{\rho^4}{4} \right]_{r_2}^{r_1}$$

$$M = \frac{2 \pi q_s}{r_1} \cdot \frac{(r_1^4 - r_2^4)}{4}$$

$$M = \frac{\pi q_s}{2 r_1} (r_1^4 - r_2^4) = \frac{\pi q_s}{16 d_1} (d_1^4 - d_2^4).$$

91. The advantage of the hollow shaft over the solid is that the strength is greater for the same weight of material. With shafts of the same length, the weight varies in proportion to the cross-section areas. These are  $\pi r^2$  for the solid and  $\pi (r_1^2 - r_2^2)$  for the hollow shaft. From the mathematical theory the hollow shaft should be a thin shell. In practice the inner radius is made about one-half the outer. Assuming  $r_2 = \frac{1}{2} r_1$ ,  $\pi (r_1^2 - r_2^2) = \pi (r_1^2 - \frac{1}{4} r_1^2) = \frac{3}{4} \pi r_1^2$ . Making this equal to  $\pi r^2$  of the solid shaft so as to have the two of the same weight,  $\frac{3}{4} \pi r_1^2 = \pi r^2$  or  $r_1^2 = \frac{4}{3} r^2$ . Then, substituting the value of  $r_1$  in terms of  $r$  in the strength formulas above, the strength of the hollow shaft compares with the strength of the solid shaft of the same weight as

$$\frac{\pi q_s \frac{16}{16} \cdot \frac{16}{9} r^4}{2 \cdot \frac{2}{\sqrt{3}} r} \text{ is to } \frac{\pi q_s r^3}{2} \text{ or as 1.443 is to 1.0.}$$

It does not pay to make small shafting hollow unless light weight is of extreme importance. In the case of very large shafting, such as the propeller shaft of a steamship, it is impossible to forge the material satisfactorily throughout the cross-section when the piece is solid. For the sake of thorough working during the forging, a hollow piece may be used, forging it down upon a mandrel. This not only improves the quality of the material, but also takes advantage of the higher strength of the hollow shafting of the same weight. •

92. The formulas for the relation of stress intensity to external moment, derived above, are no longer true as soon as the elastic limit is passed. They give an indication of the way in which the piece will fail. They show that the stress intensity varies through a cross-section from zero at the center to a maximum at the surface. It follows that the outer parts of the piece must pass their elastic limit or break before the inner parts. **Failure of any kind, permanent deformation or break, must originate in the surface of the piece and progress inward.**

93. To find the real relation between stress intensities in different parts of the material and the external moment, after elastic action has ceased, we must return to the key equation,  $\tan \delta = \frac{\alpha}{L} \rho$ , and the general relationship of  $q$  to  $\tan \delta$ . The key equation is true whether conditions are elastic or not.

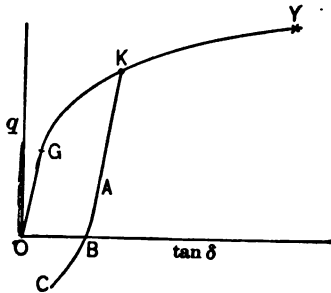


FIG. 35.

Elastic action brings in the **special relation** that  $q$  is proportional to  $\tan \delta$ . The **general relation** of  $q$  to  $\tan \delta$  is shown in Fig. 35. This relation we know by general experience of the relation of stress intensity to deformation. There is usually an elastic range  $OG$  followed by  $GKY$  where permanent deformations come in. The curve terminates in break at  $Y$ .

94. Fig. 36 shows the distribution of deformation and stress intensity from center to surface of the piece at two stages of the loading. Fig. 36A shows the relations while conditions are elastic. The magnitude, but not the directions, of  $\tan \delta$  and  $q$  are plotted vertically

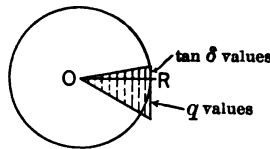


FIG. 36A.

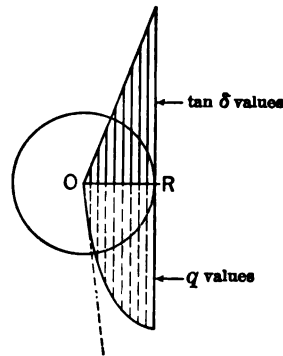


FIG. 36B.

from the radial base line  $OR$ ,  $\tan \delta$  above and  $q$  below. For elastic loading the magnitudes of both  $\tan \delta$  and  $q$  vary from center to surface,  $O$  to  $R$ , in direct proportion to the radius. Fig. 36B shows conditions near the breaking load for a ductile metal.  $\tan \delta$  values still



vary in direct proportion to radius from center to surface, but the magnitudes of the values are very much larger than during elastic action. The elastic limit value of  $\tan \delta$  is always less than 0.005. The breaking value of  $\tan \delta$  in ductile metals may be as high as 2.0. The  $q$  values in Fig. 36B are larger than under the elastic conditions, but by no means increased in the same proportions as the  $\tan \delta$  values. The variation of  $q$  from the center to surface is a curve of the same form as that of Fig. 35. It is substantially the curve of Fig. 35 compressed laterally to fit it onto the base  $OR$ .

95. In the engineering handling of the torsion test, we measure a series of simultaneous values of torsion moment  $M$  and angle of twist  $\alpha$ . From these and the dimensions of the piece we compute  $q$  in the outer surface by the formula  $q_s = \frac{2}{\pi r^3} \cdot M$  and deformation in the outer surface by the formula  $\tan \delta_s = \frac{r}{L} \cdot \alpha$ . (The  $q_s$  formula assumes the customary solid round piece.)

The  $\tan \delta$  formula can be used with satisfactory accuracy throughout the test because  $r$  and  $L$  remain nearly constant.  $r$  actually decreases and  $L$  increases so that the  $\frac{r}{L}$  ratio decreases. The changes in  $r$  and  $L$  are of the magnitude in each case of about one per cent for very ductile materials and are much less with lower ductility of the material.

With the  $q_s$  formula the case is far different; it cannot be used with satisfactory accuracy beyond the elastic limit. The departure from elastic action is large enough so that, even with brittle materials, impossible values of  $q_s$  are computed toward the end of the test. **The values of  $q_s$  obtained by the elastic formula outside of its range of validity are purely nominal. They are greater than the real values.** The discrepancy between these nominal values and the real values increases as the loading goes on. We shall denote these nominal values of  $q_s$  by  $q_n$ .

96. From a test we can determine the relation of  $q_n$  to  $\tan \delta$ , for any given material. We cannot get directly from the test measurements of a torsion test the real relation of  $q$  to  $\tan \delta$ , or shear stress to deformation, for that material. But since the relation of  $q$  to  $\tan \delta$

determines the form of the curve of  $q_{ns}$  against  $\tan \delta_s$ , an indirect solution is possible.

Let the relation of  $q$  to  $\tan \delta$  be the general function  $q = \phi(\tan \delta)$ . Then  $q_s$ , the real shear stress in the surface, is  $\phi(\tan \delta_s)$ , and  $\tan \delta_s$  can be truly found at all stages of the loading from  $r$ ,  $L$  and  $\alpha$ . As before, take as a unit for mathematical analysis a ring of material in the cross-section, of width  $d\rho$ , at radius  $\rho$ , Fig. 34. The area of this ring is  $2\pi\rho \cdot d\rho$ .  $q = \phi(\tan \delta)$  is constant around this ring. The total shear tangent to the ring is  $q \cdot (\text{area of ring}) = q \cdot 2\pi\rho \cdot d\rho$ . The moment arm of this total shear is  $\rho$ .

Hence

$$dM = \rho \cdot q \cdot 2\pi\rho \cdot d\rho = 2\pi q \rho^2 d\rho.$$

$\rho$  must be put in terms of  $\tan \delta$ . Using the key equation  $\tan \delta = \frac{\alpha}{L} \cdot \rho$ , we put first

$$\tan \delta_s = \frac{\alpha}{L} \cdot r.$$

Hence the proportion

$$\frac{\tan \delta}{\rho} = \frac{\tan \delta_s}{r} \quad \text{or} \quad \rho = \frac{r}{\tan \delta_s} \cdot \tan \delta.$$

Then

$$d\rho = \frac{r}{\tan^2 \delta_s} \cdot d(\tan \delta).$$

Substituting, in the equation for  $dM$ , the values of  $\rho$  and  $d\rho$  just derived in terms of  $\tan \delta$  and  $d(\tan \delta)$ , we have the general differential equation

$$dM = 2\pi q \cdot \left( \frac{r}{\tan \delta_s} \right)^2 \cdot (\tan \delta)^2 \cdot \frac{r}{\tan^2 \delta_s} \cdot d(\tan \delta).$$

Collecting terms,

$$dM = 2\pi \left( \frac{r}{\tan \delta_s} \right)^3 \cdot q \cdot (\tan \delta)^2 \cdot d(\tan \delta).$$

Putting  $\phi(\tan \delta)$  for  $q$ ,

$$dM = 2\pi \left( \frac{r}{\tan \delta_s} \right)^3 \cdot \phi(\tan \delta) \cdot (\tan \delta)^2 \cdot d(\tan \delta).$$

Integrating, bringing constant terms outside of the integral sign,

$$M = 2\pi \left( \frac{r}{\tan \delta_s} \right)^3 \cdot \int_0^{\tan \delta_s} \phi(\tan \delta) \cdot (\tan \delta)^2 \cdot d(\tan \delta).$$

Now  $q_{ns}$ , the nominal stress intensity in the outer surface by the elastic formula, is  $\frac{2}{\pi r^2} \cdot M$ .

$$\begin{aligned} \text{Hence} \quad q_{ns} &= \left( \frac{2}{\pi r^2} \right) \cdot 2\pi \left( \frac{r}{\tan \delta_s} \right)^3 \cdot \int_0^{\tan \delta_s} \phi(\tan \delta) \cdot (\tan \delta)^2 \cdot d(\tan \delta) \\ &= \frac{4}{\tan^3 \delta_s} \cdot \int_0^{\tan \delta_s} \phi(\tan \delta) \cdot \tan^2 \delta \cdot d(\tan \delta). \end{aligned}$$

Put  $U$  for  $\frac{4}{\tan^3 \delta_s}$  and  $V$  for the  $\int_0^{\tan \delta_s} \phi(\tan \delta) \cdot \tan^2 \delta \cdot d(\tan \delta)$ .

Then  $q_{ns} = U \cdot V$ . Differentiate this with regard to  $\tan \delta_s$ , to get the slope of the curve of  $q_{ns}$  versus  $\tan \delta_s$ , at the abscissa  $\tan \delta_s$ . This curve is the one determined experimentally and known from the test.

$$\begin{aligned} \frac{d}{d(\tan \delta_s)} (q_{ns}) &= U \frac{d}{d(\tan \delta_s)} (V) + V \frac{d}{d(\tan \delta_s)} (U) \\ &= \frac{4}{\tan^3 \delta_s} \cdot \frac{d}{d(\tan \delta_s)} \left( \int_0^{\tan \delta_s} \phi(\tan \delta) \cdot \tan^2 \delta \cdot d(\tan \delta) \right) \\ &\quad + \left( \int_0^{\tan \delta_s} \phi(\tan \delta) \cdot \tan^2 \delta \cdot d(\tan \delta) \right) \frac{d}{d(\tan \delta_s)} \left( \frac{4}{\tan^3 \delta_s} \right). \end{aligned}$$

In the first term of this expression the value of

$$\frac{d}{d(\tan \delta_s)} \left( \int_0^{\tan \delta_s} \phi(\tan \delta) \cdot \tan^2 \delta \cdot d(\tan \delta) \right)$$

is the differential of an integral with regard to its upper limit and has the value  $\phi(\tan \delta_s) \cdot \tan^2 \delta_s$ . For this to be true it is necessary that  $\phi(\tan \delta)$  be a continuous finite single-valued function of  $\tan \delta$ . We know that  $q$  is such a function of  $\tan \delta$ . In the second term of the differential of  $q_{ns}$ , the value of

$$V = \left( \int_0^{\tan \delta_s} \phi(\tan \delta) \cdot \tan^2 \delta \cdot d(\tan \delta) \right) \text{ is } q_{ns} \cdot \frac{\tan^3 \delta_s}{4}.$$

Making these substitutions we have

$$\begin{aligned} \frac{d}{d(\tan \delta_s)} (q_{ns}) &= \frac{4}{\tan^3 \delta_s} (\phi(\tan \delta_s) \cdot \tan^2 \delta_s) + \frac{q_{ns} \cdot \tan^2 \delta_s}{4} \left( \frac{4(-3)}{\tan^4 \delta_s} \right) \\ &= \frac{4 \cdot \phi(\tan \delta_s)}{\tan \delta_s} - \frac{3 q_{ns}}{\tan \delta_s}. \end{aligned}$$

But  $\phi(\tan \delta_s)$  is the real value of  $q_s$ . Hence

$$\frac{d}{d(\tan \delta_s)} (q_{ns}) = \frac{4 q_s}{\tan \delta_s} - \frac{3 q_{ns}}{\tan \delta_s}.$$

Multiplying through by  $\tan \delta_s$ , we obtain

$$\tan \delta_s \cdot \frac{d}{d(\tan \delta_s)} (q_{ns}) = 4 q_s - 3 q_{ns}.$$

Change signs, and add  $q_{ns}$  to each side.

$$\text{Then } \left\{ q_{ns} - \tan \delta_s \frac{d}{d(\tan \delta_s)} (q_{ns}) \right\} = 4 (q_{ns} - q_s).$$

A tangible meaning may now be given to the mathematics above. Graphically, on the plot of  $q_{ns}$  vs.  $\tan \delta_s$ , the left-hand side of the last equation, viz.  $\left\{ q_{ns} - \tan \delta_s \frac{d}{d(\tan \delta_s)} (q_{ns}) \right\}$ , is the intercept on the  $q$  axis of the tangent to the curve at the point of which the abscissa is  $\tan \delta_s$ .

Fig. 37 shows, in the full line curve  $OEA$ , the experimental curve of  $q_{ns}$  vs.  $\tan \delta_s$ . At any value  $D$  of  $\tan \delta_s$ , the ordinate  $DA$  cuts the curve at  $A$ . From  $A$ ,  $AB$  is drawn tangent to the curve, cutting the  $q$  axis at  $B$ .  $OB$  is the value of

$$\left\{ q_{ns} - \tan \delta_s \cdot \frac{d}{d(\tan \delta_s)} (q_{ns}) \right\}$$

for  $\tan \delta_s = OD$ .

In the same equation, the expression  $(q_{ns} - q_s)$  is the error of the

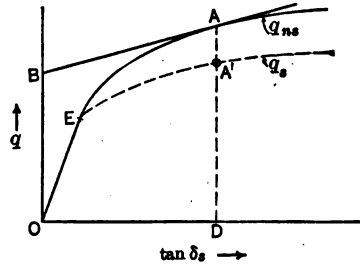


FIG. 37.

nominal value of  $q$ , or is the difference between the  $q_{ns}$  value calculated by the elastic formula and the real value of  $q_s$  in the surface. To locate the real value of  $q_s$ , lay off downward on  $AD$ , Fig. 37,  $AA'$  equal to  $\frac{1}{4}$  of  $OB$ .  $DA'$  is then the true value of  $q_s$  for the abscissa  $\tan \delta_s = OD$ . By repeating the construction the entire curve of real values of  $q_s$  vs.  $\tan \delta_s$  may be found point by point from the experimentally determined curve of  $q_{ns}$  vs.  $\tan \delta_s$ .

Throughout the mathematics above, there has been no assumption whatever as to the form of the function  $q = \phi(\tan \delta)$  except the valid one that it is a continuous, finite, single-valued function. The graphical construction thus proved for the finding of the real value of  $q$  is therefore valid for any material, brittle or ductile, at any stage of the loading. The proof applies, however, only to a solid round shaft.

97. In commercial testing the construction for finding the real value of  $q_s$  is particularly useful at two stages of the test. These places are: first, the neighborhood of the yield point, where the con-

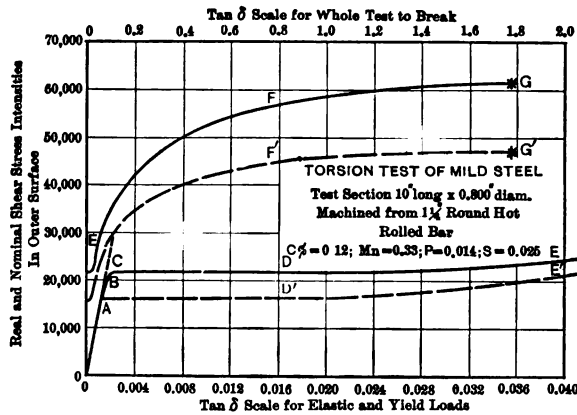


FIG. 38.

struction is of decided help in finding the true yield-point stress; and second, at the break, where by means of the construction we find from the torsion test the real breaking strength of the material in shear. (The form of the break must be noted to be sure that the piece did break in shear, not in tension.)

Fig. 38 shows the curves obtained from careful test of a mild steel. Two scalings are used for the abscissas,  $\tan \delta_s$ , to bring out in the lower

curves the relations around the yield point and in the upper curves the relations at break. The full lined curves are the nominal stress values and the dashed curves are the real stress values.  $A$  is the elastic limit. The real stress follows  $OAD'E'$ . The nominal stress corresponding is  $OACDE$ . It might be possible by accurate observation to pick out, on the line  $OAC$ , the point  $A$  where there is the first slight departure from the straight line. The chances, however, are very much against this. If the construction to find the curve of real stress  $OAD'E'$  were not made, one would almost surely pick the point  $C$  as the yield-point load, making an overestimate of between 20 and 25 per cent. The apparent analogy of  $OACDE$ , the nominal stress curve, with the similar flat after the yield point in tension testing, would seem to justify the taking of  $C$  as the yield-point stress. This example should make plain that **the application of the construction in finding of the real stress intensity is necessary in determination of shear yield-point stress intensity in the torsion test.** The true yield point is at  $A$ , not at  $C$ .

98. At break, the fact that  $q_{nsB}$  is not the real breaking strength of the material in shear has long been recognized in practice by naming  $q_{nsB}$  "**the modulus of rupture.**" In other words, the modulus of rupture is  $\frac{2}{\pi r^3} \cdot M_B$  (the subscript  $B$  referring to break). In Fig. 38,  $G$  is the break point on the nominal stress curve and the stress intensity at  $G$  is the modulus of rupture. The point  $G'$  at the end of the real stress curve gives the real breaking strength of the material in shear. **Again we would not find the real shear breaking strength from the torsion test if we did not have the construction for getting from the nominal stresses to the real stresses.**

99. The value of the "**modulus of rupture**" is a property of a material. It is to be used in calculating the torsion strength of a piece of material from its dimensions. The torsion strength of a round solid shaft at break is  $M_B = \frac{\pi r^3}{2} \cdot q_{nsB}$ . This use of "modulus of rupture" for torsion strength calculation corresponds to the use of the "engineering" maximum tension stress times area of original cross-section, in calculating the tension strength of a piece. When a material is used in direct shear, as for example, a rivet, we may not use the "mod-

ulus of rupture" to calculate the breaking strength of the piece, but must use the real shear breaking strength of the material,  $q_{sB}$ .

100. The more ductile the material, the flatter is the end of the nominal stress-deformation curve. Hence the more ductile the material the more nearly does the real shear breaking strength of the material,  $q_{sB}$ , approach to  $\frac{3}{4}$  of the value of the modulus of rupture,  $q_{nsB}$ . Fig. 39 is an illustration of this. Tests were made of a series of materials in both torsion and direct shear. The torsion tests gave moduli of rupture and values of  $\tan \delta_s$  at break for the materials. The

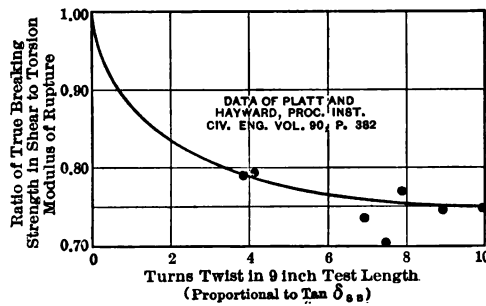


FIG. 39.

direct shear tests gave approximately the real shear breaking strengths. Fig. 39 is a plot of the ratio of real shear breaking strength to modulus of rupture for each material vs. the values of  $\tan \delta_s$  at break for that material. The ratio  $\frac{q_{sB}}{q_{nsB}}$  must lie between 1 and  $\frac{3}{4}$ , and rapidly approaches  $\frac{3}{4}$  as ductility or  $\tan \delta_{sB}$  increases. Fig. 39 shows that for most of the ductile metals we will not make serious mistakes by assuming the ratio  $\frac{q_{sB}}{q_{nsB}}$  to have a value of 0.80 to 0.75.

101. Fig. 40 shows the common method of making the direct shear test, to determine the real breaking strength in shear. A pin  $XY$  of the material to be tested is made a close fit to a hole through three hard steel plates. The two outer plates are pulled in one direction, the inner plate in the opposite direction, in a tension testing machine. The pin is sheared at both sides of the middle plate. The trouble with this testing method is that if the plates do not fit closely together, the piece is bent as well as sheared; while if the plates do fit closely



Fig. 29, from the edge of the break of Fig. 41. If diagonal planes  $BFHD$  and  $AEGC$  are passed through this cube, it will be seen that the shear forces  $Q_1, Q_2, Q_3, Q_4$ , combining in pairs, set up a tension stress across one of these diagonal planes,  $BFHD$ , and a compression stress across the other diagonal plane,  $AEGC$ . The compression stress may be neglected because it cannot cause trouble, but the tension stress may cause a break. Fig. 42B shows the relations of

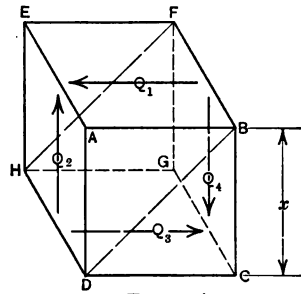


FIG. 42A.

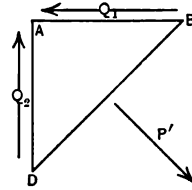


FIG. 42B.

the total shears  $Q_1$  and  $Q_2$  on two faces of the cube and the total tension force  $P'$  on the diagonal plane.  $P' = 2 Q \cos 45^\circ = 2 Q \frac{\sqrt{2}}{2}$ .

The intensity of shear stress on the face of the cube is  $q = \frac{Q}{x^2}$ . All the  $Q$ 's are equal and  $x$  is the length of one edge of the cube. The area of the diagonal plane is  $x \cdot x \sqrt{2} = x^2 \sqrt{2}$ . The tension stress intensity on the diagonal plane is

$$p' = \frac{P'}{x^2 \sqrt{2}} = \frac{2 Q \frac{\sqrt{2}}{2}}{x^2 \sqrt{2}} = \frac{Q}{x^2} = q.$$

By a little more mathematical work, it may be shown that this secondary tension stress,  $p' = q$ , is the maximum value of the tension stress.

**105.** Since the brittle materials break in the secondary tension stress and not by shear, it is impossible by torsion loading to find the shear strengths of brittle materials. The "modulus of rupture" of brittle materials in torsion corresponds closely to their breaking strengths in tension. The modulus of rupture is higher than the



tension strength, because the elastic formula for the torsion loading is not true at the break point of the brittle materials. As we have seen, the nominal stresses in torsion loading are higher than the real stresses. The ratio of real tension strength to the torsion modulus of rupture is about  $0.8 \pm$  for the brittle metals.

**106.** After a piece has been loaded in torsion beyond its elastic limit, the outer parts of the piece, having higher stresses than the inner parts, have permanent deformation, while the inner parts are still elastic. Taking off the external load then does not reduce the internal stresses to zero. Fig. 36B shows the general nature of the internal stress distribution under load, beyond the elastic limit. In Fig. 35, after a portion of the material has been stressed to the point *K*, unloading gives a stress-deformation relation *KABC* (a "set line"), not *KGO*. From the stress distribution curve of Fig. 36B each concentric layer of material in the cross-section of the piece unloads along such a set line of its own. The result may be shown in two stages. In Fig. 43A, the outer surface of the test piece has reached zero stress intensity in the unloading. The inner parts have had relatively less

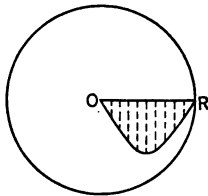


FIG. 43A.

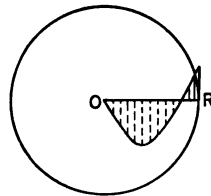


FIG. 43B.

reduction of deformation and stress intensity so that they still have stresses in the original direction. The unloading of the piece as a whole is not yet complete. When the external load is entirely removed from the piece, the internal stress condition is shown in Fig. 43B. In the inner parts of the piece there still remain stresses in the original direction, less than before. The outer parts of the piece have their stresses reversed from the original direction. They have been forced to some such condition of stress and deformation as indicated by the point *C*, Fig. 35. The moment in the original direction, due to the stresses in the inner parts of the piece, is balanced by an opposite moment due to the reversed stresses in the outer parts of the piece.

The only way in which internal stresses can be removed from a piece which has been loaded in torsion beyond the elastic limit is by an annealing heat treatment.

**107.** In gripping a test piece for torsion loading, the important condition for accurate work is that the grips apply to the piece pure torsion loading uncomplicated by tension, compression or bending. The gripping heads must be supported by some form of universal joint to avoid bending moments. At least one grip must be free for endwise motion, to avoid tension and compression. Squared ends on the test piece, though often specified, are unnecessary when the grips of the testing machine are mounted in universal joints. Bending moment on the test piece causes a minor oscillation of load superposed on the torsion load. The diameter of the test section of the piece must be less than the diameter in the grips to prevent premature break.

**108.** While the solid (or hollow) round cross-section is by far the most common shape for pieces carrying torsion loading, there is occasional use of the square, rectangular or hexagonal shapes. There are also complicated shapes which are due to single or multiple keyways or splines. With these it is probably safe to consider the effective size of the shaft as measured to the root of the keyway or spline. A key and keyway will not reduce the strength of a shaft provided that the following conditions are met: shaft must be a driving fit in the hub; key must be a driving fit in both shaft and hub; the key must completely fill the keyways in length, breadth and depth; the keyway must not extend along the shaft outside of the hub on the side where the shaft carries torsion moment, in other words the keyway of the shaft must stop within the hub.

**109.** For a solid round shaft, of diameter  $d$ , with elastic loading, we have found that  $q_s = \frac{5.09}{d^3} \cdot M$ , or  $M = 0.1966 q_s d^3$ . The derivation of the corresponding formula for a square shaft is too complicated to give here. The result is  $q_s = \frac{4.8}{b^3} \cdot M$ , or  $M = 0.2083 q_s b^3$ ,  $b$  being the side of the square. In elastic loading at the same stress, the relative strengths of the square and round shafts are as 1.058 to 1, when  $b = d$ . If the diagonal of the square shaft equals the diameter

of the round, as when the end of a round shaft is milled or forged down to square,  $b = 0.707 d$  and the relative strengths of the square and round shafts are as 0.375 to 1 for elastic loading at the same stress. The relative breaking strengths of round and square shafts of ductile material can be approximated by assuming a uniform stress intensity through the cross-section equal to the shear breaking strength of the material. Then the strengths of round and square are  $M = 0.262 q_B d^3$  and  $M = 0.405 q_B b^3$ . If  $b$  equals  $d$ , the side of the square equals the diameter of the round, the breaking strength of the square is 1.545 times the breaking strength of the round. For  $b = 0.707 d$ , the diagonal of the square equal to the diameter of the round, the breaking strength of the square is 0.546 times the breaking strength of the round.

These relative strengths, elastic and breaking, show that when putting a squared end on a round shaft, economy of material, or use of the full strength of the round, demand that the side of the square should equal the diameter of the round. Furthermore, the change of cross-section from one size or shape to another must always be done through rounded transition forms. Sharp corners, especially re-entrant ones, must be avoided. Good practice demands generous fillets.

**110.** The materials used for shafting are ductile metals, usually steels. Line shafting in mills is of cold rolled steel because the cold rolling process both improves the strength of the material and finishes the material very closely to exact sizes with an excellent surface. Crank shafts of engines are forgings. In large steam and gas engines the crank-shaft material is wrought iron or mild steel forged and machined. In automobile engines, the crank-shaft material is medium carbon steel or alloy steel, forged, machined and heat treated.

## CHAPTER VI

### TRANSVERSE LOADING

**111. Transverse loading means any system of forces which causes bending of the piece.** Transverse loading is usually made up of forces or components of forces perpendicular to the length of the piece in planes passing through the lengthwise axis of the piece. Most commonly the forces are due to gravity and lie in vertical planes. A beam is an example of this vertical transverse loading.

**112. The measure of a transverse load is the bending moment,** the product of a force and a lever arm. Bending moment is usually variable along the length of the piece. **The value of the bending moment at any cross-section of a beam is the algebraic summation of products of forces times lever arms, taken to one side of the section.**

The lever arms are to be measured from the center of the section to the line of action of each force.

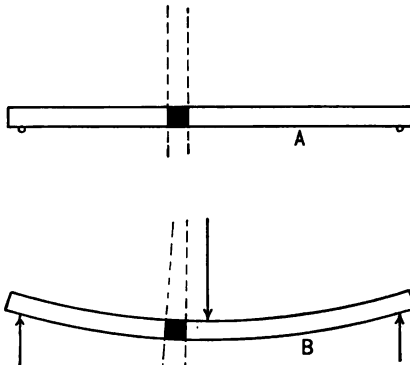


FIG. 44.

**113. The external deformation is a bending of the piece.** Fig. 44B shows the relation of the loaded beam to the same beam without load, Fig. 44A.

**The measure of the bending is either the deflection of any part of the beam from its initial unloaded position, in a direction**

**perpendicular to the length of the piece; or, the radius of curvature at a portion of the piece.** The deflection measurement is much the more common one.

**114. The internal deformation of the material can be arrived at by isolating for study a small portion of the length of the piece.** In Fig. 44A the vertical dashed lines represent the edges of two planes

perpendicular to the length of the piece. When the beam is not loaded these planes are parallel and the portion of the beam between them is rectangular from a side view. When the beam is loaded (Fig. 44B) these planes, remaining perpendicular to the beam, are no longer parallel, but have become so inclined to each other that they intersect at the center of curvature for the portion of the beam between them. The change of shape, from the side view, of the portion of the beam between the two planes, is exaggerated in Fig. 45. *A* represents the unloaded, *B* the loaded, condition. In *A*, a layer of material parallel to the length of the beam is of the same length at any place from top to bottom of the part. In *B*, the different layers vary in length from top to bottom. **The internal deformation of the material is therefore a change of length in a direction parallel to the length of the piece. This internal change of length varies across the beam, from the upper surface to the**

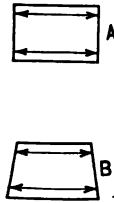


FIG. 45.

**lower, in direct proportion to distance from the surfaces. This linear variation of the internal deformation with distance from the bent surfaces is the key to the mathematical theory of the transverse loading. The variation of the deformation remains of the same kind, whether the loading is elastic or not. When the bending moment varies along the length of the piece the linear law of variation of the internal deformation across the section is still true, but is masked by a second kind of deformation (shear) superposed upon it.**

**115. The internal stresses corresponding to a change of length of a material are tensions or compressions. Hence, from the nature of the internal deformations, there must be set up by transverse loading tension or compression stresses, or both, in directions parallel to the length of the piece. The internal stresses cannot be all tension or all compression, because then the length of the entire piece would change. No change of length of the piece as a whole can occur, because in the external loading there are no forces parallel to the length of the piece. Hence the internal tensions on a given cross-section must balance in amount the compressions on that same section. Another way of proving this necessity of balance of the tension and compression stresses on a given section is shown in Fig. 46. A portion of the beam at one end is taken as a free**

body. At the cut end,  $XY$ , the action of the material removed is replaced by the internal stresses by which the part removed was acting on the part remaining. There being no external forces parallel to the length of the piece, equilibrium against endwise motion of the part in Fig. 46 requires a balance of the tension and compression stress amounts on the section  $XY$ .

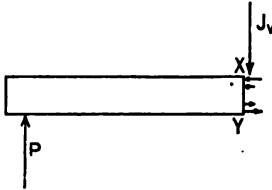


FIG. 46.

116. Since the internal lengthwise deformations vary in a straight-line law through the beam, the tension and compression stresses, which are proportional to the deformations in elastic loading, must, while the loading is elastic, follow the same kind of straight-line law. The two conditions, equality in amount of total tension and total compression on a section, and variation of stress intensity according to a straight-line law across the section, lead to the conclusion that during elastic loading the zero point for both stresses and deformations must be at the center of gravity of the cross-section. More generally, there is a plane through the piece parallel to the inner and outer bent surfaces and passing along the length of the piece through the centers of gravity of successive sections, a plane of zero stress and deformation. This is termed the **neutral plane**. The intersection of the neutral plane with any cross-section plane is the **neutral axis** of that cross-section. From the neutral axis to the inner bent surface of the piece the internal stresses are compressions, increasing in intensity in proportion to distance from the neutral axis. Similarly, from the neutral axis to the outer bent surface of the piece the internal stresses are tensions, increasing in intensity in proportion to the distance from the neutral axis. Fig. 47 shows the distribution of deformations and stress intensities between the inner and outer bent surfaces while the loading is elastic.  $BE$  is the edge of a cross-section plane. The deformation distribution is shown by  $ABOEF$ , the stress distribution by  $CBOED$ .

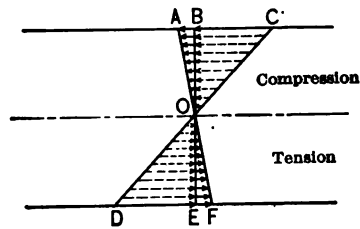


FIG. 47.

117. We are now able to derive the formula connecting the maximum stress intensities of compression and tension, in the surfaces of the beam, with the shape and size of the cross-section and the magnitude of the external moment at the section. Fig. 48 shows at the left a cross-section through a beam, at the right the variation of stress intensity through this cross-section. The neutral axis need not be

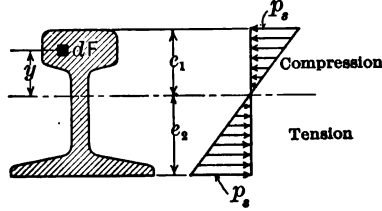


FIG. 48.

halfway between the upper and lower surfaces of a beam, for the cross-section need not be symmetric about the neutral axis. The distances  $e_1$  and  $e_2$ , from neutral axis to surfaces of the beam need not be equal. When  $e_1$  and  $e_2$  are not equal, the surface stress intensities in tension and compression are not equal. Consider an element of area  $dF$  of the cross-section at a distance  $y$  from the neutral axis. The tension or compression stress intensity acting on  $dF$  is  $p$ . The force  $p \cdot dF$  has a moment about the neutral axis equal to  $y \cdot p \cdot dF$ . The stress intensity desired is  $p_s$ , the value in the outer surface of the beam. Hence  $p$  must be expressed in terms of  $p_s$ . Since stress intensity varies in proportion to distance from the neutral axis  $\frac{p}{p_s} = \frac{y}{e}$ . Then

$$p = \frac{y}{e} \cdot p_s. \quad dM = y \cdot p \cdot dF = y \cdot \frac{y}{e} \cdot p_s \cdot dF = \frac{p_s}{e} \cdot y^2 \cdot dF.$$

Integrating, 
$$M = \int dM = \frac{p_s}{e} \cdot \int_{-e_2}^{+e_1} y^2 dF.$$

$M$  is the total moment of the internal stresses on the cross-section. It is also the moment of the external forces with regard to that section.  $e_1$  and  $e_2$  are distances from the center of gravity of the cross-section to the surfaces. The  $\int_{-e_2}^{+e_1} y^2 dF$ , the summation of product of each element of area by the square of its distance from the neutral axis, is called the **moment of inertia** of the cross-section. It is denoted by  $I$ . The formula becomes  $M = \frac{p_s I}{e}$ , or  $p_s = \frac{Me}{I}$ .

118. The study of the safety of a given case of transverse loading resolves itself generally into finding the location and magnitudes of

## CONSTRUCTION

The cross-section of a beam is given, and we have to find the position of the neutral axis and then solve  $p_s = \frac{Me}{I}$ . The beam must for correct design be made of a certain material. If the weight of the beam  $p$ , must be found, the curves plotted for its variation with length, a graphical solution may be found. The beam occurs as a practical problem, keeping  $p$ , constant, and finding the length of the beam.

It is evident that material is wasted in a beam of constant cross-section, if the bending moments in proportion to the distance from the neutral axis. A beam of rectangular cross-section, as in Fig. 49A, is wasteful of material. Hence in practice, material is often cut away from the neutral axis is often cut away, as shown in Fig. 49B. This is the so-called "I-beam" section. The upper and lower parts of this section are called "flanges." The central connecting portion is the "web."

The figure shows the I-beam as against the rectangular section, Fig. 49A, and the ratio of  $\frac{I}{e}$ , "section modulus,"

$$F = 6 \times 2 = 12.$$

$$F = 4.5.$$



Hence

$$\frac{I}{eF} = \frac{20.38}{3 \times 4.5} = 1.512.$$

The I-beam in this case is one and one-half times as strong per unit weight as the rectangular sectioned beam. The actual strength of the solid rectangular section of the same overall dimensions is, however, the greater in this case in the ratio 1.767 to 1.

**120. Transverse loading causes not only primary tension and compression stresses, but primary shears as well.** This may be shown by Fig. 46, representing one end of a beam as a free body. The beam has been cut at  $XY$ , on which the tension and compression stresses are indicated. The tension and compression stresses on  $XY$  are not sufficient to represent all of the stresses which must be acting on that section. Beyond  $XY$  at the left is the vertical force  $P$  on the end of the beam. To balance this, for vertical equilibrium, there must be a total vertical shear force  $J_v$  across the section  $XY$ . In general the total vertical shear across any section of a beam is equal to the algebraic summation of all the vertical loading forces to one side of that section. The vertical shear force is not uniformly spread over the cross-section of the beam to give a constant stress intensity at all parts of the section. The law of variation of shear stress intensity across the section must be investigated.

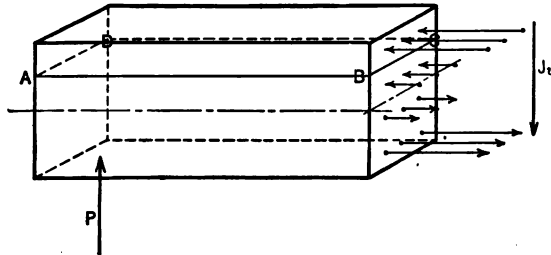


FIG. 50.

**121.** Fig. 50 is a perspective view of the same piece of material shown in Fig. 46.

Through the beam near its top is passed a plane  $ABCD$ . The portion of the beam above  $ABCD$  is to be lifted off as a new free body shown in

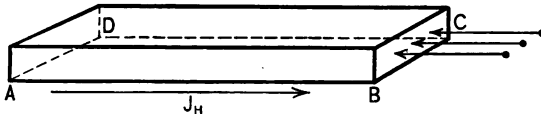


FIG. 51.

Fig. 51. On the right-hand end  $BC$  of this new free body we have a set of the internal compression stresses of the beam acting as end thrusts. There being

no external forces on the beam to balance these end thrusts, we discover that there must be, for horizontal equilibrium, a horizontal shear  $J_H$  on the plane  $ABCD$ .

**122.** We have now shown that there are both horizontal and vertical shear stresses within the beam. In Fig. 30 and its discussion, Section 79, we have previously shown that shear stresses are always paired in planes at right angles to each other and that the shear stress intensities at any given point are equal in the two directions. Hence at any given point in a beam the intensities of the vertical and horizontal shear stresses are identical. It follows that the distribution of stress intensity on a cross-section is the same for the horizontal and vertical shears, for we can make a complete point by point substitution of one for the other.

**123.** The law of distribution on a cross-section for the vertical shear stresses is difficult to come at directly. The distribution for the horizontal shear stress intensities is comparatively easy to derive. The plan of analysis is the free body method of Figs. 50 and 51.

For complete generality the free body studied must be cut out from the beam by two cross-section planes  $UV$  and  $XY$  at a distance  $dx$  from each other (Fig. 52). The horizontal analysis plane, cutting

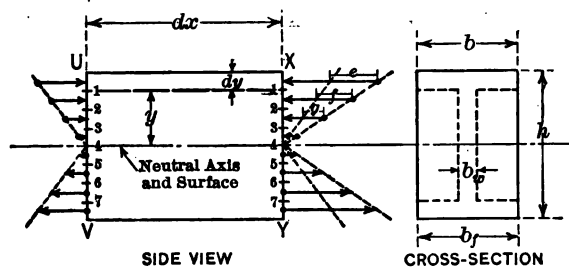


FIG. 52.

off a portion of the top of the beam, is made to take successively positions 1-1, 2-2, etc., from top to bottom of the beam. As the analysis plane moves through these positions it will reveal the variations on itself of its total horizontal shear and of the intensity of that shear. The external moments acting on the cross-section planes  $UV$  and  $XY$  are in general different, so that the tension and compression stresses on  $UV$  and  $XY$  will be of different magnitudes. The differ-

ence of magnitudes of the tension and compression stresses on  $UV$  and  $XY$ , at the various levels through the beam, are indicated by  $e, f, g$ , etc. Since on both  $UV$  and  $XY$  the tension and compression stress intensities vary in direct proportion to the distance from the neutral axis, the differences  $e, f, g$ , etc. must also vary in direct proportion to the distance from the neutral axis. Placing the analysis plane at 1-1, the end thrust due to unbalance of the compression stresses at  $U$  and  $X$  is  $e \cdot b_1 \cdot dy$ .  $b_1 dy$  is the area of the strip of the cross-section above the plane 1-1 on which the stress difference  $e$  is acting. At the level 1-1  $J_H$  is then equal to  $e \cdot b_1 \cdot dy$ . The area of the plane 1-1 is  $b_1 \cdot dx$ . Therefore  $q_H$ , the intensity of the horizontal shear on the plane 1-1, is  $\frac{e \cdot b_1 \cdot dy}{b_1 \cdot dx}$ . Let the analysis plane now be moved to the position 2-2. Then the unbalanced total end thrust, or the horizontal shear  $J_H$ , is the sum of the previous value  $e \cdot b_1 \cdot dy$  plus the amount picked up in the change from 1-1 to 2-2, which is  $f \cdot b_2 \cdot dy$ . The area of the plane 2-2 is  $b_2 \cdot dx$ .  $q_H$  on the plane 2-2 is then  $\frac{e \cdot b_1 \cdot dy + f \cdot b_2 \cdot dy}{b_2 \cdot dx}$ .

In general,

$$q_H = \frac{\int_{(\text{analysis level})}^{(\text{top of beam})} (p_u - p_z) \cdot b dy}{b_{(\text{analysis level})} \cdot dx} = \frac{\int_y^e \Delta p \cdot b dy}{b \cdot dx}.$$

Since  $p = \frac{p_z}{e} \cdot y$ , and  $p_z = \frac{Me}{I}$ ,

$$p = \frac{Me}{eI} \cdot y = \frac{M}{I} \cdot y \text{ and } \Delta p = \frac{dM}{I} \cdot y.$$

Hence 
$$q_H = \frac{\int_y^e \frac{dM}{I} y \cdot b dy}{b \cdot dx} = \frac{1}{Ib} \cdot \frac{dM}{dx} \cdot \int_y^e b y dy.$$

It may be noted that in this formula  $\frac{dM}{dx} = J_v$  at a section. Hence the formula may also be written  $q_H = \frac{1}{Ib} \cdot J_v \cdot \int_y^e b \cdot y \cdot dy$ .

**124.** There are two special forms of cross-section for which we should study the distribution of shear stress intensity. They are the solid rectangular section and the I-beam section. The channel section may be considered an I-section with the web at one side in-

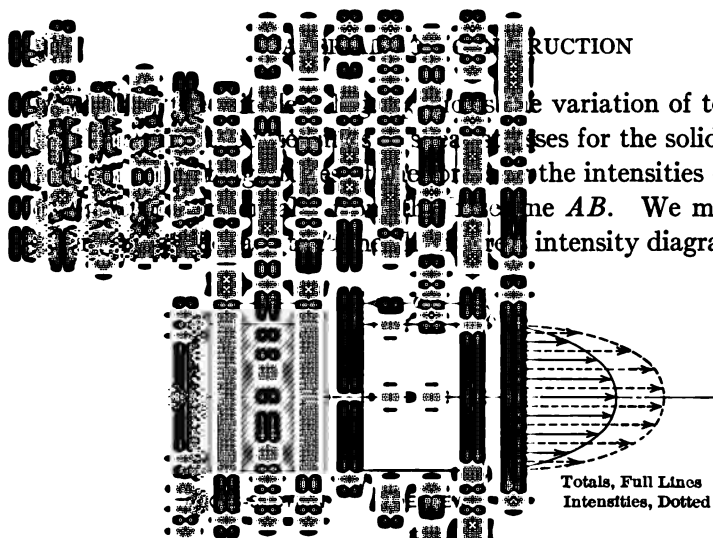


FIGURE 1

the variation of total horizontal stresses for the solid rectangular cross-section. The intensities of the shears are shown in the diagram AB. We must now call these intensity diagrams serve for

shows, since the two are everywhere equal, vertical or horizontal, the maximum at the top and bottom of the cross-section the curve of variation of the intensity of a parabola with its vertex at the center of the cross-section. The intensity of the vertical (or horizontal) shear stress is  $J_v \div (\text{area of cross-section})$ . The average shear stresses is

$$J_v = \frac{1}{2} \cdot b \cdot \int_0^h y dy = \frac{J_v}{I} \left[ \frac{y^2}{2} \right]_0^h = \frac{3}{2} \cdot \frac{J_v}{bh}$$

the maximum shear stress intensity is  $\frac{3}{2}$  times the average. It is to be noted that the loading is more uniform in the primary shear stresses than in the primary tension and compression stresses. Wood

is very weak in shear along the grain. The horizontal shear stresses in a wooden beam may be dangerous, although their numerical values seem not comparable with the tension and compression stress intensities.

**126.** The I-beam section has nearly all of its tension and compression forces on the flanges. Consequently, the total horizontal shear jumps to large values in crossing the flanges and is nearly constant through the height of the web. This is shown in Fig. 54. The

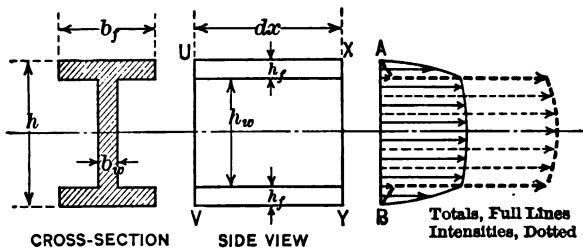


FIG. 54.

width of the I-beam is large in the flanges and small in the web. Hence the shear stress intensity is comparatively small in the flanges and large in the web, jumping suddenly in passing from flanges to web. **We can use in practice the approximation that the entire vertical shear is carried by the web of the I-beam with a uniform distribution.** This assumption is probably within 5 per cent of the truth in the ordinary case. It makes the maximum horizontal or vertical shear stress intensity  $q_{\max} = \frac{J_v}{h_w b_w}$ . (See Fig. 54 for the dimensions  $h_w$  and  $b_w$ .)

**127.** We have now studied two sets of primary stresses, a tension-compression set, and the shear set of horizontal and vertical shears. There is a possible interaction of the two sets. We may replace the tension and compression stresses by their equivalent set of secondary shears. These lie at 45 degrees to the tension and compression directions and have intensities numerically one-half as great as the tensions and compressions. The dangerous stress in the case of ductile metals, which fail in shear, and in the case of wood, is the resultant shear stress between the primary horizontal or vertical shears and the

secondary shears which replace the tensions and compressions. The important case in practice, where this resultant shear must be provided for, is at the junction of web and flanges in I-beam or channel sections.

When at a given point in the material there is acting a primary tension or compression stress of intensity  $p$  and at the same point there is acting a primary shear of intensity  $q$  parallel and perpendicular to the tension or compression, the maximum resultant shear stress lies at an angle  $\theta$  with the direction of the tension or compression given by  $\tan 2\theta = -\frac{p}{2q}$  and has the intensity  $q'_{\max} = \sqrt{p^2 + 4q^2}$ .

As an example, suppose an I-beam to be designed to carry simultaneously tension and compression stress in the flanges of 40,000 pounds per square inch, equal to the tension or compression yield points, and horizontal and vertical shears in the web equal to 20,000 pounds per square inch, equal to the shear yield point. Such a design would result from the ordinary assumptions of the stress duties of the different parts of an I-beam and the attempt to secure maximum economy of material. If the shape of the I-beam cross-section were that given in Fig. 54 the resultant dangerous shear stress at the junction of web and flanges would be

$$q'_{\max} = \sqrt{p^2 + 4q^2} = 22,350 \text{ pounds per square inch.}$$

This would cause failure of the web of the I-beam where it joins the flanges. To avoid such failure it is necessary to thicken the web at the junction with the flanges, using a generous fillet. It is a happy accident that an I-beam cannot be rolled without having such a fillet. In the actual I-beam the extreme outer layers of the flanges and the central part of the web are the danger spots.

128. Actual cases of loading in engineering practice are intended to be always well under the yield point of the material. The formulas which have so far been derived for transverse loading have all depended upon elastic action. When the piece becomes permanently deformed or breaks, the action is not elastic and formulas based on the assumption of elastic action are therefore untrue. If the "elastic" formulas of transverse loading are applied where the conditions are not elastic the calculated stresses are higher than the real stresses. In order to understand the failure of the material and in order to

make intelligent provision in practical design against failure we must know more or less accurately what really goes on in the material after elastic action has ceased. We must know not only the direction of the error of our formulas but also the magnitude of their error.

129. The basic fact of the transverse loading which we can still use, after the elastic limit is passed, is the linear law of the variation of the internal lengthwise deformation at a cross-section, from one to the other of the bent surfaces. From tension and compression tests of the material we will know the "physical" stress-deformation characteristic curves of the material. Types of these curves are shown in Figs. 3, 8, 20, 21. Fig. 3 is general for either tension or compression. Fig. 8 is the tension curve of cast iron. The compression curve of cast iron is similar in form to the tension curve but is much more nearly elastic and runs to a strength in compression which is approximately four times that of tension. For ductile metals the compression curves, on the physical basis, are practically duplicates of the tension curves. For each value of relative elongation across a section of a piece in transverse loading, we must pick out from the proper stress-deformation curve (tension or compression) the corresponding value of stress intensity. By repeating this process the variation of stress intensity at the section studied, from one to the other of the bent surfaces of the piece, may be worked out.

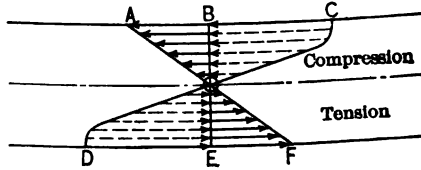


FIG. 55.

Figs. 55 and 56 show the distribution of deformation and stress intensity through the cross-section of beams of soft steel and cast iron

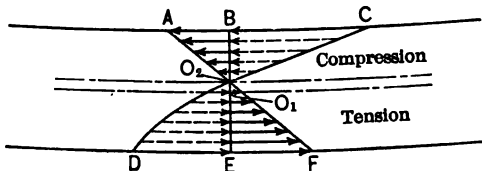


FIG. 56.

respectively at the point of loading which represents practical engineering failure. They are to be contrasted with Fig. 47, which represents the conditions during elastic action.

130. The ductile metals have practically symmetric curves of stress and deformation in tension and compression. Hence the stress dis-

tribution diagram *BCODE* (from the base *BE*, Fig. 55) is alike in top and bottom parts of the beam. If the cross-section shape of the beam is symmetric about its center of gravity, the neutral axis remains at the center of gravity though the loading goes beyond the elastic limit or yield point. Engineering failure of a beam of ductile material is not the break but the excessive permanent deflection which occurs when the yield point is passed. In an I-beam the compression flange or the web buckles.

131. The engineering failure of a beam of brittle material, as cast iron or wood, is an actual break. The brittle materials are effectively stiffer in compression than in tension. With increasing load on a beam of brittle material the neutral axis shifts toward the compression side. The stress distribution curve *BCO<sub>2</sub>DE*, Fig. 56, is not symmetric in tension and compression. The stress intensities are higher on the compression side. The break of brittle metals is a tension break on the outer bent surface of the beam. Wood may fail in a number of ways. It may break in tension on the outer bent surface. It may fail in horizontal shear along the grain near the neutral axis. It may fail in secondary shear from the high compression stresses on the inner bent surface.

132. When elastic action has ceased the ordinary formulas for the calculation of stress intensities are no longer true. The calculated tension and compression stresses are nominal values, higher than the real values. Applying the "elastic" formula  $p_e = \frac{Me}{I}$  at the breaking load of a brittle material, we calculate the **transverse "modulus of rupture."** The modulus of rupture is a property of the material, in a certain restricted sense. The modulus of rupture is independent of the size of the piece provided the character of the material does not change, but depends on the shape of the cross-section. The nature and properties of cast iron change with the size of the piece, so that it is not safe to reason from one size to another, assuming the properties constant. Cast iron is alone in this change of properties with size of piece; other engineering materials have properties nearly independent of the size of the piece. For ductile metals in I-beam sections it is safe in practice to assume that the nominal stress intensity when the piece bends so much as to have effectively failed is the same as the



tension or compression yield-point stress. For ductile metals in solid rectangular sections the nominal stress at the load where bending constitutes failure is about  $\frac{3}{2}$  times the tension yield-point stress. The use of the modulus of rupture, or the nominal stress at point of failure, is to calculate the probable load at engineering failure for any size of piece we choose.

**133.** When the compression and tension stress distribution goes to the forms shown in Figs. 55 and 56, the distribution of horizontal and vertical shear also changes. In the I-beam section the changes in the shear distribution are not important; it remains substantially as in Fig. 54. In the solid rectangular sections the shear distribution changes from a parabolic arc on the base  $AB$ , and with vertex on the neutral axis, Fig. 53, toward a triangle on the base  $AB$  with its apex on the neutral axis. That is, the ratio of maximum shear stress intensity to average shear stress intensity increases, changing from  $\frac{3}{2}$  toward  $\frac{4}{3}$  as an upper limit. The ratio probably never actually exceeds

$\frac{3.5}{2}$ . That is, the error in the use of the elastic formulas for maximum

shear stress probably does not exceed 15 per cent.

**134.** The measurements of a transverse test consist of a series of simultaneous values of loads and deflections, and the dimensions of the piece, including distances between loading forces. From the loading forces and distances between them the external moments at any section can be found.  $p_s$  is computed for the cross-section carrying the maximum external moment.

**135.** Corresponding to  $p_s$ , the unit stress intensity in outer fibers at the dangerous section, we need values of  $\epsilon_s$ , the unit deformation (relative change of length). The calculation of  $\epsilon_s$  is made from the observed deflections, and the form of curve assumed by the neutral surface of the beam along the length of the beam. Fig. 44 shows the nature of the curve taken by the beam. Formulas are derived in mechanics for the relations between the loading and the deflections at any section of the beam from the original unloaded position. These formulas have a common general form of the type deflection

$$D = \frac{K_1 \cdot P \cdot L^3}{EI}$$
, in which  $K_1$  is a constant depending on the section of beam chosen and the placing of loads,  $P$  is load,  $L$  is length of

beam,  $E$  is the tension-compression modulus of elasticity and  $I$  is the moment of inertia of the cross-section. The bending moment at any section is  $M = K_2 \cdot P \cdot L$ ,  $K_2$  being a constant similar to  $K_1$ . From the definition of modulus of elasticity we will have the relation, while the loading is elastic,  $E = \frac{p_s}{\epsilon_s}$ . Hence  $\epsilon_s = \frac{p_s}{E}$ .

In terms of the moment,

$$p_s = \frac{Me}{I} = \frac{K_2 \cdot P \cdot L \cdot e}{I}.$$

In terms of the deflection,

$$E = \frac{K_1 PL^3}{DI}.$$

Substituting these values of  $p_s$  and  $E$ .

$$\epsilon_s = \frac{p_s}{E} = \frac{K_2 PL e}{I} \cdot \frac{DI}{K_1 PL^3} = \frac{K_2 e}{K_1 L^2} \cdot D.$$

For example, with a single load  $P$  in the middle of the span of length  $L$ , the maximum moment occurs under  $P$  and has the value  $\frac{PL}{4} \left( K_2 = \frac{1}{4} \right)$ . The deflection  $D = \frac{PL^3}{48 EI} \left( K_1 = \frac{1}{48} \right)$ , also at the center of the beam. Hence  $\epsilon_s = \frac{12 e}{L^2} \cdot D$  for this case. This is the

load distribution most commonly used in testing.

**136.** When the loading passes the elastic limit this method of computing  $\epsilon_s$  fails just as the stress formulas fail. The real values of  $\epsilon_s$  are larger than the calculated values. The reason is that the bending localizes at the sections where the moment is greatest, making the deflections greater than would correspond to elastic action at the same loads.

**137.** One arrangement of the transverse loading has become practically standard for engineering practice in testing. This is the single load concentrated at the center of the span. The length  $L$  of the span is often 36 inches for wrought iron, steel and cast iron pieces of cross-sections 1 inch or 2 inches square. The transverse test with single central loading on 12 or 24 inch spans is often a specified test for cast irons. The transverse test is much more commonly used for cast irons than the tension test. For wood it is advisable

to work on as large pieces as possible. In small test pieces of wood the influence of bad spots of cross-grain or knots is much more serious than in large pieces. Hence a fairly large piece is necessary to get a representative test of the wood.

138. With single central loading the bending moments are zero over each support and increase in direct proportion to distance from the supports to the maximum at the center, under the load. This distribution of moments along the piece is the weakness of single central loading as a standard for transverse testing. It forces the failure of the material to occur at the center of the span. Any defects in the material will remain undiscovered unless they happen to be almost under the central load.

It would evidently be much better to apply a constant bending moment along a considerable length of the piece. Any defect in the material which occurs within that length will then be searched out and shown in the break. Uniform bending moment is practically as easy to obtain in testing as the variable bending moment. Two equal loads each of value  $P$  equidistant from the center of the span and each at a distance  $x$  from the support at its end of the beam will cause a uniform bending moment  $P \cdot x$  through the central portion of the beam between the two loads. Fig. 57 shows a simple testing

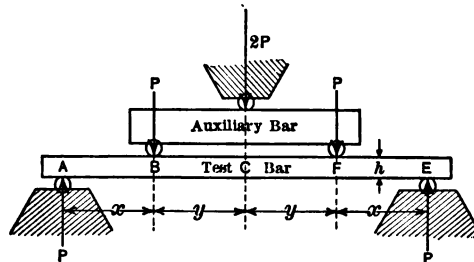


FIG. 57.

arrangement for putting the middle portion of a test bar under uniform bending moment. A further advantage of this method of loading is that if the material in the central part  $BF$  of the bar is homogeneous and uniform in cross-section this part of the bar bends into an arc of a circle whether the loading is elastic or not. Hence by measuring the deflection of the center  $C$  from the line of the sections  $B$  and  $F$

the radius of curvature and the true values of  $\epsilon_s$  can always be found for the part  $BF$  whatever the stage of the test.  $\epsilon_s = \frac{Dh}{y^2 + D^2}$  in this case,  $D$  being the deflection of section  $C$  from the line  $BF$ .

139. In Section 96 it was shown that for a solid round bar in torsion loading there is a peculiar relation between the real stress intensity in the outer surface and the nominal stress intensity. A similar relation can be proven for transverse loading on a solid rectangular section. Use a uniform bending moment on the test section, as in  $BF$  of Fig. 57, so that  $\epsilon_s$  values can be truly found at all stages of the loading. From the moments are to be calculated, by the elastic formula  $p_s = \frac{Mc}{I}$ , the nominal values of  $p_s$ . From the deflections compute the real values of  $\epsilon_s$ . Then plot the curve  $OEA$  of Fig. 58 with  $p_{ns}$  ordinates and  $\epsilon_s$  abscissas. At any point  $A$  of this curve draw the tangent  $AB$  intersect-

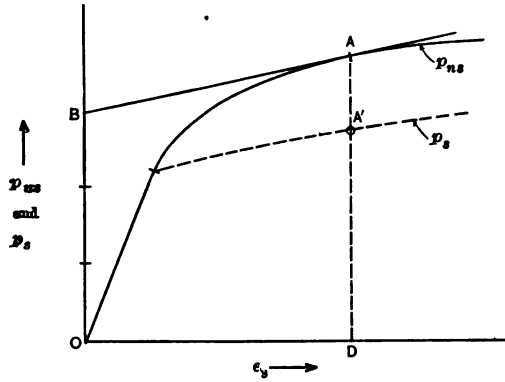


FIG. 58.

ing the ordinate for zero value of  $\epsilon_s$  at  $B$ . The error of  $p_{ns}$  at the point  $A$  is one-third of  $OB$ . On the ordinate  $DA$  lay off  $AA'$  downward from  $A$  equal to one-third of  $OB$ .  $DA'$  is then the real stress intensity  $p_s$  corresponding to  $DA$  as nominal value. By repeating the construction the entire curve of real value  $p_s$  vs.  $\epsilon_s$  can be found. This curve will be found to be the same as the "physical" curve of the tension test of the same material.

140. For the special case of bars of ductile metals in square or rectangular cross-sections, with single central loading, the true value of  $\epsilon_s$  at any stage of the loading on the main curve after the elastic limit (not on a set line) is

$$\epsilon_s = \frac{6h}{L^2} \left\{ D_E \frac{P}{P_E} + \frac{1}{3} \frac{\left( D - D_E \frac{P}{P_E} \right)}{\left( 1 - \frac{P_E}{P} \right)} \right\}.$$

In this  $h$  is the height of cross-section;  $L$  is the span between supports, in inches;  $D$  is the total deflection at the center of the beam;  $D_E$  is the center deflection at the elastic limit;  $P$  is the center load;  $P_E$  is the load at the elastic limit. This formula is to be used only after the elastic limit has been passed. Before the elastic limit use the ordinary formula,

which in this case is  $\epsilon_s = \frac{6h}{L^2} D$ . With the true values of  $\epsilon_s$  found in this way the construction method of section 139 and Fig. 58 becomes applicable for finding the true values of  $p_s$ .

141. For cast iron in transverse loading it is possible to calculate approximately the true tensile strength by a formula which differs from the ordinary elastic stress formula in that it assumes that the moduli of elasticity in compression and tension are not equal. Let  $E_c$  represent the ratio of  $\frac{p_s}{\epsilon_s}$  on the compression side of the beam and  $E_t$  a similar ratio on the tension side. We may call  $E_c$  and  $E_t$  the effective moduli of elasticity.  $E_c$  is in ordinary cast iron about twice  $E_t$ , for a range of stress as high as the tensile breaking strength. With a beam of rectangular section the tensile breaking strength is

$$T = \frac{3}{bh^2} \cdot M \left\{ 1 + \frac{1}{\sqrt{\frac{E_c}{E_t}}} \right\}.$$

The ordinary formula gives modulus of rupture  $= \frac{6}{bh^2} M$ . The modulus of rupture is considerably higher than the real tensile strength of the cast iron. The alternative formula just given is not exact, but is a much better approximation to the truth. The following table gives a comparison of the two formulas.

Ratio of effective moduli of elasticity, $\frac{E_c}{E_t}$ . . . . .	1.000	1.50	2.00	2.50	3.00
Ratio of $T$ to modulus of rupture . . . . .	1.000	0.908	0.854	0.816	0.789

## CHAPTER VII

### COMPRESSION LOADING

**142. Compression loading** consists of forces pushing on the piece. At first sight compression is the reverse of tension, and one might expect the same kind of phenomena in reversed directions. While the loading is elastic, and the stress intensity through the piece is uniform, compression and tension are the reverse of each other. When the elastic limit is passed, or when in elastic loading the stress distribution is not uniform, compression is not the reverse of tension.

Compression loading may be contrasted with tension as unstable against stable. For instance a crooked piece given a tension load tends to straighten out into the line of the loading, becoming better able to carry the loading. A crooked piece given compression loading tends to become more crooked, departing more and more widely from the line of the loading, and becoming less able to carry the loading.

**143. The failure of a piece of material under compression loading** is never due to the compression stress itself, but always to some secondary action. In short blocks of brittle material the break is ultimately in shear. Short blocks of ductile materials can hardly be said to break but flatten out instead (secondary shear action) more or less indefinitely. Long pieces in compression loading bend or buckle. Bending takes the material in the central parts of a long piece out of the line of the end loads, so superposing transverse stressing on the direct compression stressing. The final failure then is in ductile materials by secondary shear on the inside of the bend; in brittle materials either by this secondary shear on the inside of the bend, or tension break on the outside of the bend.

**144.** Pieces carrying compression loading are commonly called **struts** or **columns**. When the piece is short and stiff enough so that it does not perceptibly bend before it fails it is called a **short column**. When a piece is long enough to have perceptible bending in its failure it is a **long column**. There is no real distinction between the two

cases. The bending changes in degree only. With increase of length short columns grade into long columns without demarcation.

**145. In short columns of ductile material the only significant load is the yield-point load.** Beyond the yield point the piece takes considerable permanent deformation, becoming shorter and fatter. This permanent deformation ends the engineering usefulness of a part in a structure. Hence the yield-point stress intensity is the engineering strength of a short column of ductile material. With increasing deformation beyond the yield point the actual load on the piece increases for two reasons: because the stress intensity is rising and because the cross-section area of the piece is increasing. Materials of low ductility may be made to break in shear by increasing the loading beyond the yield point. Materials of high ductility act in a semi-plastic manner, decreasing in length and increasing in cross-section area beyond the practical limits to which the test can be carried. No true break is obtained with these.

**146.** Up to the yield point the deformation of the piece is practically all elastic deformation caused directly by the compression stresses. At and beyond the yield point the permanent deformations are caused directly by the shear stresses, which are secondary to the compression stresses. The relations of direction and magnitude of the compression and shear stresses are exactly the same as the similar relation for tension and shear derived in Sections 50 and 51 and Fig. 18, modified by reversal of stress directions. This is true only near the yield point, when the shear slippage is beginning. As the shear slippage increases within the crystals, friction comes in. The modification of the relation between shear and tension, by the friction, is not the same as the corresponding modification between shear and compression.

**The yield-point stress intensity is the same in tension and compression.** This is to be expected theoretically because the yield point is the beginning of internal shearing of the crystals in each case, with the same relation of the shear stress to the tension or compression stress. Experiment confirms the theory. The determinations of yield-point stress intensity in either tension or compression loading vary in individual tests of the same material through a range of about  $\pm 10$  per cent from the average for that material. When a

sufficient number of tests are taken for comparison the average compression yield point and the average tension yield point are identical within experimental error.

It is rather easier to make a satisfactory tension test than to make a satisfactory compression test. It follows that in engineering practice the short-column compression test of ductile material is unnecessary. We may substitute a tension test. The only information we might get from the compression test is the yield point. The tension test will give that information and more.

**147.** Short columns of brittle materials load up with scarcely perceptible deformations, either elastic or permanent, to a break which occurs without warning. This break is by shear, with the angle between the plane of the break and the normal cross-section plane through the piece somewhat greater than 45 degrees. The parts of the broken piece slip on each other, along the plane of this break, with considerable friction, because the compression load is holding the parts together.

In a ductile metal, at its shear yield-point, slip planes form through the crystals in great numbers. In a single crystal slip cannot proceed far on one slip plane before it meets a frictional resistance. The parts of the crystal across the plane of slip hold together after the slip.

It is easier to start another slip plane parallel to the first than to continue the slip on the first plane. Hence the formation of large numbers of slip planes through the crystals. In a crystal of a brittle material the parts do not hold together strongly across a slip plane. Hence the tendency is to make a complete shear break through the crystal on the first slip plane rather than to accumulate slips, as in the ductile metal.

The planes of shear break through the crystals of a piece of brittle material acting as a short column in compression loading are those natural cleavage planes of the crystals which lie nearest to and on both sides of the 45-degree plane. The failure of a few crystals transfers the load they had carried to the adjacent crystals. The break is sudden when once it has started. Since the planes of break of the individual crystals are not quite parallel from crystal to crystal the surface of break through the mass as a whole is a rough surface. The motion of the parts of the broken piece on each other, as the



motion starts, is subject to two resistances. The crystals on the path of the break must be sheared, and friction due to the compression load must be overcome.

148. Let Fig. 59A represent the piece under compression. The primary compression stress  $p = \frac{P}{F}$ , the plane of the area  $F$  being

perpendicular to  $P$ . The piece breaks along some plane  $F'$  inclined to  $F$  at the angle  $\theta$ . Fig. 59B shows the forces concerned. The component  $P \cdot \sin \theta$  of  $P$ , parallel to the plane of break, must overcome on that plane the breaking strength of the material in shear  $Q' = q_B \cdot F'$ , and the friction of motion. The force holding the parts together on the plane  $F'$  is  $P' = P \cos \theta$ . The friction is  $\mu P' = \mu \cdot P \cdot \cos \theta$ , the product of the coefficient of friction  $\mu$  and force  $P'$  perpendicular to the plane of slip.

The area  $F' = F \cdot \frac{1}{\cos \theta}$ . Hence

$$P \sin \theta = q_B \frac{F}{\cos \theta} + \mu P \cos \theta. \text{ Divid-}$$

ing by  $F$ ,  $\frac{P}{F} = p$ , the intensity of

the primary compression stress. The equation becomes

$$p \sin \theta = q_B \cdot \frac{1}{\cos \theta} + \mu \cdot p \cdot \cos \theta.$$

Solving for  $p$ ,

$$p = \frac{q_B}{\cos \theta (\sin \theta - \mu \cos \theta)}.$$

Changing to functions of the double angle,

$$p = \frac{2 q_B}{(\sin 2 \theta - \mu - \mu \cos 2 \theta)}.$$

At break  $\theta$  must have such a value as to make  $p$  as small as possible.

Therefore at the breaking load  $\frac{dp}{d\theta} = 0$ .

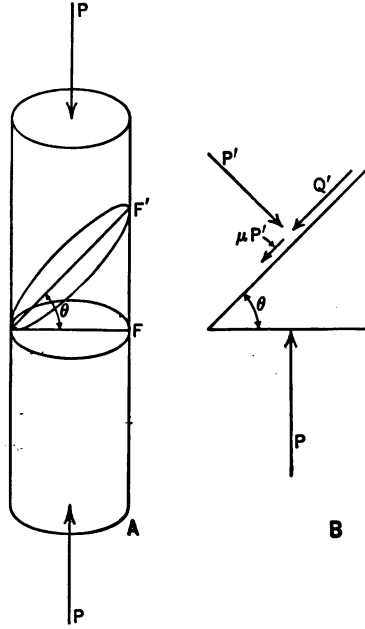


FIG. 59.

From the equation above,

$$\frac{dp}{d\theta} = \frac{-2q_B(2\cos 2\theta + 2\mu \sin 2\theta)}{(\sin 2\theta - \mu - \mu \cos 2\theta)^2}.$$

Since the denominator of this cannot be infinity, we must have

$$(2\cos 2\theta + 2\mu \sin 2\theta) = 0, \text{ or } \cot 2\theta = -\mu.$$

Let  $\mu = \tan \phi$ ,  $\phi$  being the "angle of repose" of the material for frictional slipping on itself. Then

$$\begin{aligned}\tan \phi &= -\cot 2\theta = -\tan (90^\circ - 2\theta). \\ \phi &= 2\theta - 90^\circ; \theta = 45^\circ + \frac{1}{2}\phi.\end{aligned}$$

The angle of break therefore is  $\theta_B = 45^\circ + \frac{\phi}{2}$ .

**149.** The relation of  $q_B$  to  $p_B$  is  $q_B = \frac{1}{2} p_B \cot \theta_B$ ,  $\theta_B$  having the value of  $45^\circ + \frac{\phi}{2}$ . The relations between  $q_B$ ,  $p_B$  and  $\theta_B$  for the brittle materials can be used in various ways. If the breaking strength in shear and the coefficient of friction are known, the compression breaking strength can be calculated. If the compression strength and coefficient of friction are known, the shear breaking strength can be calculated. If a compression test is carefully made so that  $\theta_B$  can be measured accurately on the broken parts the shear strength of the material can be calculated from  $p_B$  and  $\theta_B$ . The last two methods are of practical importance because a compression test is easy to make and is more accurate than a direct shear test.

**150.** Tests of short columns of either brittle or ductile materials are somewhat affected by friction of the ends of the test piece on the heads of the testing machine. The heads of the testing machine are of hardened steel much stronger than the material under test. When the short column breaks or yields, its parts tend to move outward across the heads of the testing machine. Friction of the ends of the test piece on the heads of the testing machine opposes this motion. This causes the piece to carry a higher load than it ought at the time of failure. The shorter the piece compared to its thickness the more serious is the effect of the friction against the heads of the testing machine. The error due to this cause depends also upon the coefficient of friction of the material under test with the steel heads of the

machine. Figs. 60 and 61 show the influence of shape of test piece and of the coefficient of friction in exaggerating the crushing strength of short columns. As the ratio of length to least thickness increases, a length is reached beyond which the short column tips slightly on

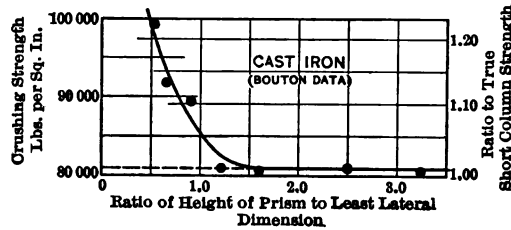


FIG. 60.

its ends at the moment of break, instead of slipping one end across the head of the testing machine. The strength is then practically independent of the length. Figs. 60 and 61 show that this critical length, which a short column must have to get a fair test, is for cast iron at

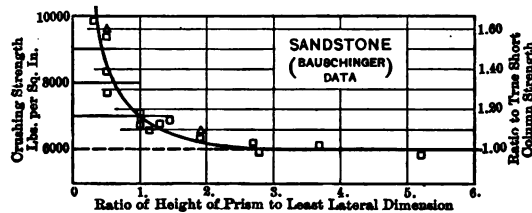


FIG. 61.

least 1.25 times the least thickness of the piece and for sandstone about 2.5 times.

151. It is an unfortunate fact that the most common form of short-column test piece, for brittle materials, especially concrete and stone, is the cube. We have just shown that this cannot give a fair test. Figs. 60 and 61 show in their ordinates on the right-hand edge the strength stated as a percentage of the true short-column strength. This shows that for sandstone the cubic test piece gives an excess strength of 18.5 per cent. This would be fairly typical of concrete and stone testing. The excess strength of a cube of cast iron is about 5 per cent. This would be typical of metals and wood. This condem-

nation of the cubic test piece for finding crushing strength applies to the use of the data so obtained as a basis of design. **The strength which can be realized in an actual structure is the crushing strength of a test piece not affected by friction from the heads of the machine.** One of the reasons for the very high factors of safety used in the design of masonry and concrete is that the factor of safety has to cover the error in strength due to using a cubic test piece. When the data from test pieces of cubic form is to be used only in comparison of strengths of similar materials the comparisons obtained will be fairly accurate. The friction on the heads of the testing machine alters the angle  $\theta_B$  of the break and may change the character of the break. The cubic form of test piece should be recognized as bad practice and should be discontinued. A proper specification would be that the length of a short column should be between three and five times its least thickness.

152. A practical case of conditions reversed from those of the too short compression test piece between the heads of the testing machine, is the cast iron or steel column thrusting against large surfaces of stone or concrete. On the bearing surface between the metal column and the stone work the effective strength of the stone work is somewhat higher than the compression strength of a test block of the stone. The reason is that the adjacent stone work furnishes lateral support to the portion of the stone receiving load from the column. Still, in order to load the metal column to its full strength it is necessary to put a base or top on it to spread the load over a considerable area of the stone work, reducing the bearing pressure to safe values for the stone work.

153. If a piece of material could be so supported laterally that the secondary shear yielding were absolutely prevented, the material would be perfectly elastic under compression stress without limit to the stress. A practical case which comes close to this is that of contact of the locomotive tire on the top of a rail. The area of contact between the tire and the rail is quite small and the pressures astonishingly large (about 150,000 pounds per square inch) although both tire and rail are elastic under the load.

154. Short-column testing requires great care to secure uniform loading of the material. Both ends of the piece must be as nearly

as possible perfect fits to the heads of the testing machine. A high corner or a high spot concentrates the loading and causes premature break. Various devices may be used to distribute the loading uniformly to make up for imperfect fitting. Such devices include the use of sheets of blotting paper or lead between the test piece and the heads of the machine, or the more elaborate schemes of allowing plaster of Paris to harden between the faces of the test piece and the heads of the machine. These devices all modify the strength of the test piece in so far as that strength is affected by friction on the heads of the testing machine.

155. Long columns differ from short ones in that they have perceptible bending. The chances are that on account of imperfect fitting at the ends, the load on a column always has some initial eccentricity. The bending increases this eccentricity. A long column then carries a double stressing, for the end thrusts set up both direct compressions and bending stresses. The bending stresses are superposed upon the direct-compression stresses. On the inside of the bend the direct-compression stress and the bending-compression stress add together. On the outside of the bend the direct-compression stress and the bending-tension stress are algebraically added, or the numerical value of the tension is subtracted from the numerical value of the compression.

156. A long column must always, because of the double stressing, be weaker than a short column of the same material and the same shape and size of cross-section. The longer the column the more easily does it bend. Therefore the strength decreases as the length increases. Stiffness against bending depends on the moment of inertia of the cross-section and the tension-compression modulus of elasticity of the material. For economic design the material in a long column must be spread out to give a large moment of inertia to the cross-section.

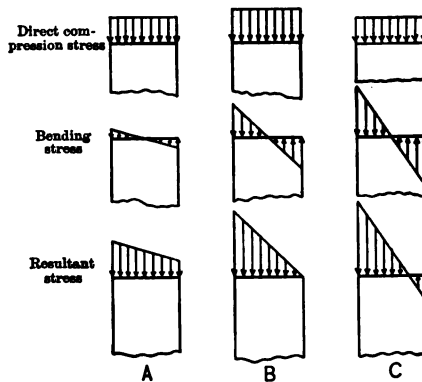


FIG. 62.

**157.** The internal stress conditions in successive stages of the loading of a fairly long column are shown in Fig. 62. In Fig. 62A a fairly large direct-compression stress is modified by a small bending stress, the resultant being a non-uniform compression stress through the piece. In Fig. 62B the load on the piece, and the direct-compression stress, have slightly increased. The bending deflection and the bending stress have considerably increased. The result is an approximate doubling of the compression stress intensity on the inside of the bend, and almost no stress at all on the outside of the bend. Fig. 62C shows the direct compression stress less than before, but the bending stresses larger. This corresponds to the conditions just before the breakdown of a fairly long column. The resultant stress intensities are a very high compression stress on the inside of the bend and a small tension stress on the outside.

**158.** In ductile materials, with the yield point at numerically the same stress for compression and tension, the final breakdown of the column is by buckling at the inside of the bend, where the numerically largest stress intensity is present. In fairly long and slender columns of brittle materials, the resultant tension stress on the outside of the bend is the cause of the final break. Stubby long columns of brittle materials finally break by the secondary shear from the compression stress on the inside of the bend.

**159.** When a long column first begins to bend its stiffness against bending is such that on removal of the load it will spring back straight. As the load increases the sidewise deflection increases at a more rapid rate than the increase of load. The action is complex. Sidewise deflection gives a moment arm to the end loads, while the resulting bending moment tends to increase the sidewise deflection. The sidewise deflection does not reach very large values before the deflection will increase at constant load, and then with decreasing loads. The load goes through a maximum value. After this maximum load the piece continues to deflect sidewise rapidly, but the load required to continue the deflection becomes less and less.

Often the material is still elastic in action after the maximum load has been passed, that is, if the load is not a gravity load, but a controlled force as in a testing machine, it is possible to carry a column beyond its maximum load and to unload with the column springing

back elastically, retracing in its unloading the curve of the loading.

160. Fig. 63 shows the relation between load and sidewise deflection of a series of columns of different lengths.  $OA$  is for a short column,

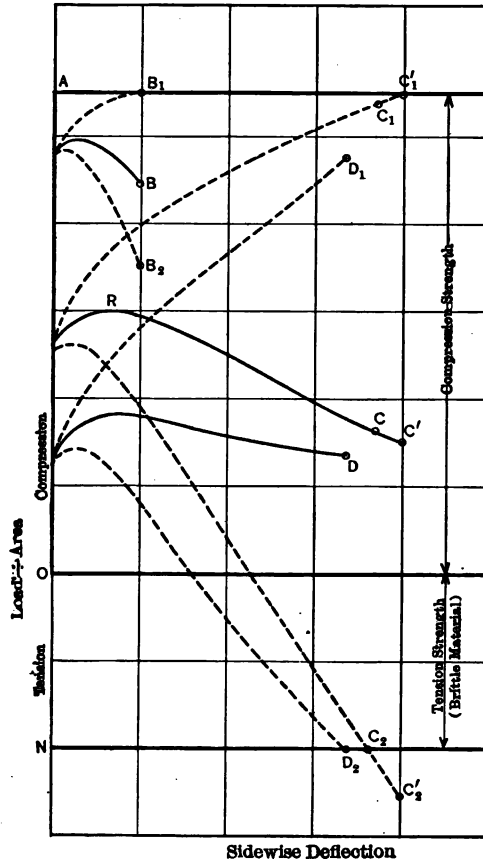


FIG. 63.

breaking without bending at a stress equal to the crushing strength of the material.  $OB$  is for a slightly longer piece bending through a maximum load to a break at  $B$ .  $OB_1$  and  $OB_2$  are the plots of the stress intensities in the inner and outer surfaces of the bent piece. Break in this piece occurs when the compression stress intensity on

the inside of the bend reaches, at  $B_1$ , the crushing strength of the material.

The stress plotted in the full line  $OB$  is the nominal or average compression stress for the whole column and is equal to actual load divided by the area of cross-section, or  $\frac{P}{F}$ . The bending stress equals the bending moment times the distance from neutral axis to outer fibers, divided by moment of inertia of the section, or  $p = \frac{Me}{I}$ . By measuring the deflections at successive loads in Fig. 63, the bending moment, and hence the bending stress, can be computed. Then the maximum stress intensity in the piece is the sum of the direct-compression stress and the bending stress; the minimum stress intensity is the difference between the two.

$ORC$ , Fig. 63, is the curve for a still longer column, for which the bending stresses are quite as important as the direct-compression stresses. If this piece failed by shear due to the maximum stress reaching the limiting short-column value at  $C'_1$ , the nominal stress at failure would be  $C'$ . Such would always be the case for ductile material. The tension strength of the brittle materials is considerably lower than the compression strength. Assume that the column giving the curve  $OC$  is of a brittle material with a tension strength  $ON$ . Then the break point will not be  $C'$  due to the maximum stress reaching  $C'_1$  but  $C$  due to the minimum stress reaching its limiting value  $C_2$ . The point  $C$  is nearer to  $R$  than is  $C'$ . We find then that when a long column of brittle material breaks by tension the actual load causing the break is higher, nearer to maximum load  $R$ , than if the break had been due to compression (shear) failure on the inside of the bend. All pieces of the brittle material longer than the piece giving the curve  $ORC$  will break by tension.  $OD$  is an example of this.

**161. The engineering strength of a long column is the maximum load the column will carry.** Engineering loads in practice are commonly gravity loads. A load less than the maximum will hardly even start sidewise deflection and is safe. A load as large as the maximum will not only start deflections, but will be more than able to continue the deflections and finish the piece. The engineering formulas for the strength of long columns should therefore aim to calculate the maximum load which the column will carry. The actual magnitude of the breaking load of a column is to a designer a matter of indifference, because the column cannot get to its breaking load without going through its maximum load.



**162.** An important factor in the strength of a long column is the **end condition**. When the end of a column is built into the adjacent material so that the column is rigidly held at the end and cannot deflect, the condition is called **fixed ended**. Steel columns riveted at their ends to girders or built into masonry are examples of fixed ends. So also is the piston rod of a steam engine. When the end of the column is cut off square and butted against the adjacent material, the column is **square ended**, **butt ended** or **flat ended**. The name "square ended" for such a column may be misleading because the column can tip, shifting the load to one edge and so is not forced to be square with the adjacent material, as a fixed-ended column must be. Stone and wooden columns in ordinary construction are examples of butt-ended columns. If the end of a column receives load from a pin or hinge the column is **pin ended**. A connecting rod in an engine is an example. In the plane of the axis of the pin such a column is fixed ended. It is free to bend only in the plane perpendicular to the axis of the pin. A column free to bend in any direction at an end is called **round ended**. A derrick boom is an example of a round-ended column.

The end condition affects the strength of a column by limiting in some way the sidewise deflection of the column. Fig. 64 shows the nature of the relation of sidewise deflection to end condition of the column. The simplest case is the pin-ended column, Fig. 64C. The bent column takes the general shape of a portion of a sine curve with the points of inflection at the pins. For a given length of column and load the deflection of the pin-ended column is larger than that for any other end condition. The fixed-ended column, Fig. 64A, can bend only in its central portion. This portion becomes a sine curve with the points of inflection *R* and *S* halfway between the center and ends of the column. The portion *RS* of the fixed-ended column is then com-

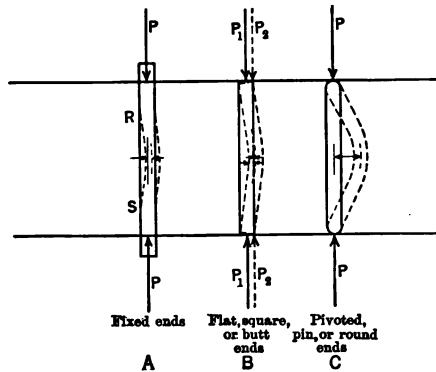


FIG. 64.

parable with the entire length of the round-ended column. With similar curvatures the deflections vary as the square of the length of the curve. Hence the sidewise deflection of the fixed-ended column is  $(\frac{1}{2})^2$  or  $\frac{1}{4}$  that of the comparable pin-ended column. The butt-ended column, Fig. 64B, does not act exactly like either of the others. With small loads a butt-ended column imitates in its deflection curve a fixed-ended column. As the loading and deflection increase, the ends of the butt-ended column tip up and the column then approximates to the curvature of a pin-ended column. In one important fact the butt-ended column differs from both the pin-ended and the fixed-ended. In neither of these does the line of action of the loading shift as the piece bends. With the butt-ended column, when the ends tip, the line of action of the loads shifts to the high corners, from  $P_1-P_1$  to  $P_2-P_2$ , as indicated in Fig. 64B. This shift makes the effective eccentricity of the loading less than the apparent eccentricity. The sidewise deflection of a butt-ended column is between that of the fixed-ended and that of the round-ended. If the length of the column is not great compared with its thickness a butt-ended column is practically fixed ended. As the length increases the effective action of a butt-ended column approaches more nearly that of the pin-ended column.

163. The following symbols will be used for writing the long column formulas:

$P$  = load in pounds;  $P_M$  = maximum load;  $P_B$  = break load.

$F$  = area of cross-section.

$K$  = radius of gyration of section.

$I$  = moment of inertia of the cross-section.

$L$  = free length of column.

$C$  = engineering crushing strength of the material in short columns. For ductile materials  $C$  = yield-point stress; for brittle materials  $C$  = break stress.

$m$  = coefficient of relative sidewise deflection, as controlled by end conditions of the column. For fixed ends  $m = 1$ ; for round or pin ends  $m = 4$ ; for flat ends  $m = 2 \pm$ . (Values of  $m$  are a matter of definition.)

$w = F \div K^2$  = a constant depending on the shape of the cross-section.

The principal formulas are then:

$$P_M = \frac{F \cdot 4 \pi^2 E}{m \cdot \frac{L^2}{K^2}} = \frac{4 \pi^2 EI}{m L^2} \quad \dots \quad \text{(Euler)}$$

$$P_M = FC \cdot \left( 1 - \frac{mC}{16 \pi^2 E} \cdot \frac{L^2}{K^2} \right) \quad \dots \quad \text{(J. B. Johnson)}$$

$$P_B = \frac{FC}{1 + \frac{mC}{4 \pi^2 E} \cdot \frac{L^2}{K^2}} \quad \dots \quad \text{(Ritter-Merriman)}$$

$$P_B = \frac{FC}{1 + m\beta \frac{L^2}{K^2}} \quad \dots \quad \text{(Rankine-Gordon)}$$

The Euler formula tries to find what end force will cause a bending moment which will make the ratio of (increase of deflection)  $\div$  (increase of load) equal to infinity. It tries to calculate the point *R* of the curve *ORC*, Fig. 63. As it neglects the effect of direct compression, and deals only with the bending moments as a possible cause of failure, the Euler formula is far too high for the usual shapes of columns. For very long columns,  $\frac{L}{K} > 300$  for steel, or  $> 80$  for wood, the Euler formula holds very well, and is a satisfactory design formula.

The Johnson formula is confessedly an empirical transition formula to change from  $P_M = FC$ , true for short columns, to tangency with the Euler curve at the point where the strength has fallen to one-half that of a short column of the same material and cross-section. Despite its empirical nature the Johnson formula, used in its proper range, fits the maximum loads (engineering strength) with surprising accuracy. It is a good design formula for long columns whose probable strength  $P_M$  is greater than one-half  $FC$ .

The Rankine formula may be regarded as an embryonic stage of the Ritter. Both attempt to compute the breaking load, on the assumption that break (or buckling) occurs when the maximum internal stress reaches the value *C*. The Rankine formula entirely disregards the fact that the breaking load of a column is not the same as its maximum load. By adjusting the constant  $\beta$  the Rankine formula

may be made to fit  $P_M$  at any chosen value of  $\frac{L}{K}$ ; but it will then calculate too low for  $P_M$  at smaller values of  $\frac{L}{K}$  and too high at larger values. As the latter kind of error is fatal,  $\beta$  of the Rankine formula is practically the same as  $\frac{C}{4\pi^2 E}$ , and the formula becomes identical with the Ritter formula.

The Ritter formula starts with the same assumption as the Rankine, and evaluates the coefficient  $\beta$  into  $\frac{C}{4\pi^2 E}$  by equating the Rankine and Euler formulas for  $\frac{L}{K} = \text{infinity}$ . For finite lengths  $\frac{C}{4\pi^2 E}$  is too large and the calculated breaking strength is less than the actual. As the designer wants the maximum strength and not the breaking strength there is not much excuse for the use of either the Rankine or Ritter formulas.

**164.** For comparison of the various long-column formulas a basis must be found which eliminates the nature of the material and the end condition of the column and states the maximum or engineering strength of the column as a function of the shape. This mathematical basis is obtained by plotting  $\frac{P_M}{FC}$  as ordinates against  $\frac{mC}{4\pi^2 E} \cdot \frac{L^2}{K^2}$  as abscissas.  $\frac{P_M}{FC}$  is the ratio of the engineering strength of the column to the strength of a short column of the same cross-section and material, taken as unity. It was pointed out before in Section 162 and Fig. 64 that the coefficient  $m$  is for the purpose of reducing columns of different end conditions to the same effective length. Hence the term  $\frac{mC}{4\pi^2 E} \cdot \frac{L^2}{K^2}$  puts all columns on comparable basis so far as the interactions of end conditions, strength and stiffness of material, length, and shape and size of cross-section, are concerned. Fig. 65 plots the engineering strengths (maximum loads) of a great number of columns of different shapes, sizes and materials on the general basis of comparison. It shows that the Johnson formula is accurate for  $\frac{P_M}{FC}$  between 1.00 and 0.50; for  $\frac{P_M}{FC}$  below 0.50 the Johnson formula

is too low while the Euler formula is accurate. The curve of the Ritter formula is drawn in to show that it does not apply at all to maximum or engineering strengths. The Ritter formula is always under the engineering strength and therefore at least safe. The ratio of  $P_M$  to calculated  $P_B$  by the Ritter formula is in ordinary design shapes around 1.3 +; this ratio has a maximum value of 1.56. The use of

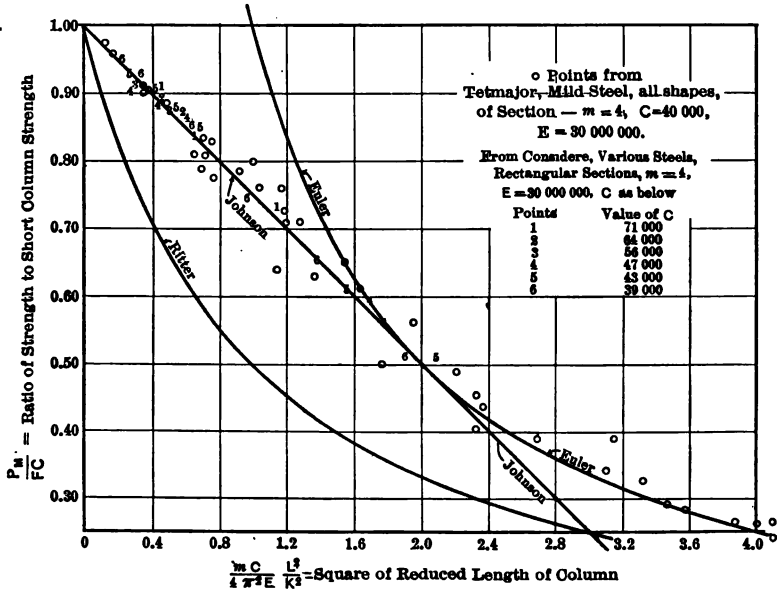


FIG. 65.

the Johnson formula, on a column design longer than that for which the formula is true, would lead to an over-large column, a safe design, though wasteful and perhaps absurd. The use of the Euler formula on too short a column would be a fatal mistake, for it vastly underestimates the necessary amount of material in the shorter columns.

The choice of a design formula depends on two things: (1) accuracy; (2) ease of calculation. Equating factor of safety times working load to  $P$  of the formulas, and solving for  $F$ , we obtain:

$$F = \sqrt{\frac{mwPL^2}{4\pi^2E}} = \frac{P}{C} \sqrt{\frac{mwC^2L^2}{4P\pi^2E}} \quad \dots \quad (\text{Euler})$$

$$F = \frac{P}{C} \left( \frac{1}{2} + \frac{1}{2} \cdot \sqrt{1 + \frac{mwC^2L^2}{P\pi^2E}} \right) \quad \dots \quad (\text{Ritter})$$

$$F = \frac{P}{C} + \frac{mwCL^2}{16\pi^2 E} \quad \dots \dots \dots \text{(Johnson)}$$

The Johnson formula evidently yields the most simple and straightforward solution; Euler next, and Ritter last. If  $\frac{mC}{16\pi^2 E} \cdot \frac{L^2}{K^2}$  for a column design equals or is less than  $\frac{1}{2}$ , the Johnson formula is true; if this discriminant is greater than  $\frac{1}{2}$ , the Euler formula is true. As this checking of the validity of the Johnson or Euler formulas is exceedingly simple, and these formulas are both easier to use than the Ritter, and have the advantage of accuracy against uncertain (though safe) inaccuracy, there seems no real reason other than the inertia of practice for the survival of the Ritter (and Rankine) long-column formula in engineering work. In routine design the designer would know without the need of checking whether to use the Johnson or

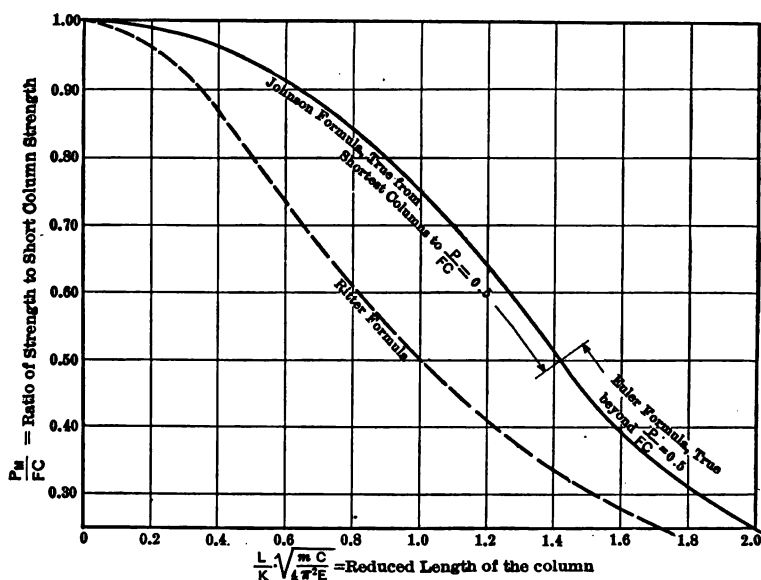


FIG. 66.

Euler formula; for example, steam-engine piston rods would always come under the Johnson formula.

165. Fig. 66 is a plot of the same quantities represented in Fig. 65, with the change of the abscissas to proportionality with  $\frac{L}{K}$  instead of

$\frac{L^2}{K^2}$ . The way in which strength varies with  $\frac{L}{K}$  is much more easily visible in this plot than in the other. It brings out also the relations of "short columns" and "long columns." There is no real dividing line. For small values of  $\frac{L}{K}$  the strength is practically independent of the length, as indicated by the flatness of the curve at its beginning. Using  $\frac{L}{K} \sqrt{\frac{mC}{4\pi^2 E}}$  instead of  $\frac{L}{K}$  in stating the length of a column reduces columns of all shapes and all materials to a common basis of comparison, so that all would plot on the full-line curves of Fig. 66.

## CHAPTER VIII

### CROSS-RELATIONSHIPS OF LOADINGS AND COMBINED LOADINGS

166. In the discussion in the preceding chapters it has been shown that in the various kinds of loading, tension, torsion, transverse and compression, failure of the material results from one or the other of two kinds of stress, tension or shear. Whatever the loading we must analyze the internal stressing to find the position and magnitude of the maximum tension stress, primary or secondary, and of the maximum shear stress, primary or secondary.

167. **The yield-point failure of ductile material occurs when the maximum shear stress intensity reaches a definite value which is characteristic of the particular material.** Since the beginning of shear slippage, at the yield point, is practically free from friction on the slip planes, the relation between the shear stress intensity at yield point,  $q_E$ , and the primary loading stress is a simple one. For tension or compression  $q_E = \frac{1}{2} p_E$ . Torsion loading is calculated directly in shear as the primary stress. Great caution must be used in torsion loading in finding the yield point. Since the failure takes place in the outer fibers first, while the inner parts of the piece are still elastic, there is not in torsion loading the sharp indication of the yield point which is obtained in tension or compression. The apparent yield-point stress of a torsion piece is not the true yield point. This is shown in Section 97. In transverse loading  $q_E = \frac{1}{2} p_E$ , but in the finding of  $p_E$  there is trouble similar to that in the torsion test. Section 139 has shown how the real values of  $p$  can be found from the nominal values in transverse loading. Dealing always with real values of stress, not nominal values, it will be found that the relation  $q_E = \frac{1}{2} p_E$  of yield point caused by shear to yield point apparently caused by normal stress, holds throughout all the systems of loading for the ductile materials.



168. Beyond the yield point the shear slippage in the crystals of ductile materials is accompanied by friction on the slip planes. The amount of this friction depends on the amount of the slip which has taken place and on the kind of loading. Engineering research has not yet gone far enough to evaluate the cross-relationships of stress and deformation beyond the yield point, between tension, compression, transverse and torsion loadings. When at some future day these cross-relationships between the "physical" values of stresses and deformations in the various loadings are worked out, it will be possible to deduce from the "physical curve" for any one loading the corresponding curves for any other loadings of that material. Perhaps we might better say that we will be able to calculate one and the same "physical curve" from any system of loading. **A statement of properties of a ductile material in terms of a tension test as a standard is satisfactory because the properties of the material in any other kind of loading are implicitly related to those in tension.**

169. The ductile materials finally break by shear. Approximately the stress relationship at break is  $q_B = \frac{1}{2} p_B$ . The tension or compression stress intensity  $p_B$  must be stated in its "physical" value. More accurately the relation from tests seems to be  $q_B = 0.42 p_B$ , the change of the coefficient being due partly to friction on the slip planes and partly to errors of assumption as to directions of stress action.

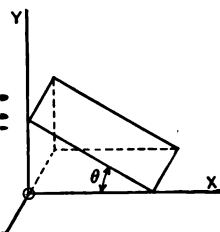
170. The ductile materials are very strong against pure tension break and relatively weak against shear yielding or shear break. The brittle materials are comparatively weak against tension break. **We might classify the materials by the relations between their tension and shear strengths. If the tension stress for pure tension break exceeds  $2 q_B$ , which in turn exceeds  $2 q_E$ , the material is fully ductile. If the tension stress for pure tension break is less than  $2 q_B$  but exceeds  $2 q_E$  the material is semi-ductile. It will have a yield point and take appreciable permanent deformation before break, but the form of the break will be like that of a brittle material. If the tension stress for pure tension break is less than either  $2 q_B$  or  $2 q_E$ , the material is brittle. It will have no yield point in tension loading and no appreciable permanent deformation before break. There is a possibility that a material brittle in tension loading might have a yield point and be somewhat ductile in compression loading. In the case**

# INSTRUCTION

if  $q_B$  is less than  $q_E$ , that is, break by pure tension when they do that they break by shear. without warning.

Obviously more than one kind of torsion, it is said to be under failure remain the same as calculations become more complex occurs in ductile materials than in brittle materials at break. be examined. The maximum sustained loading must be safely the maximum resultant tension material is brittle. Semi-ductile

combined loading choose for the resultant stresses are likely combined torsion and transverse stresses are in the surface of the piece, For irregularly formed pieces



have to be investigated. Com- produced at the analysis point others were absent. Then putting and that certain directions of loadings. For instance, one of the directions is parallel to the length of the

piece; another of the loadings may cause another tension (or a compression) also parallel to the length of the piece. These two tensions may then be numerically added. After making such combinations there will result at the analysis point a known condition of stressing such as is represented for the point  $O$ , Fig. 67. The directions  $OX$ ,  $OY$ ,  $OZ$ , for the action of the stresses, will generally be parallel to the dimensions of the piece or the directions of the external forces. Fig. 67 shows the most general case probable, having at the point  $O$  both normal stresses and shears on two planes at right angles to each other. These planes are arbitrarily taken as vertical and horizontal in Fig. 67, although they might be in any direction in the actual material. The shear stress intensities must be identical in the two directions. The tension or compression stress intensities in the two directions may have any relationship to each other and to the shears. By convention a tension stress is positive and a compression stress negative.

Pass an analysis plane near the point  $O$ , Fig. 67, at an infinitesimal distance from  $O$ . The problem is to find the values  $p'$  and  $q'$  of the normal and shear stress intensities on this analysis plane, and then by varying the angle  $\theta$  to find the maximum possible values of  $p'$  and  $q'$ .

Let the area of the inclined analysis plane be  $F'$ . Putting the summation of horizontal components of forces equal to zero, we obtain

$$p' F' \sin \theta + q' F' \cos \theta - p_x F' \sin \theta - q F' \cos \theta = 0.$$

From summation of vertical components equal zero,

$$p' F' \cos \theta - q' F' \sin \theta - p_x F' \cos \theta - q F' \sin \theta = 0.$$

Dividing by  $F'$  the two equations become

$$p' \sin \theta + q' \cos \theta - p_x \sin \theta - q \cos \theta = 0.$$

and

$$p' \cos \theta - q' \sin \theta - p_x \cos \theta - q \sin \theta = 0.$$

Eliminating  $q'$  between the two equations, and solving for  $p'$ ,

$$p' = p_x \sin^2 \theta + p_y \cos^2 \theta + 2 q \sin \theta \cos \theta.$$

Changing to functions of the double angle by the relations

$$\sin^2 \theta = \frac{1 - \cos 2\theta}{2}, \quad \cos^2 \theta = \frac{1 + \cos 2\theta}{2}$$

and

$$2 \sin \theta \cos \theta = \sin 2\theta,$$

$$p' = \frac{p_x + p_y}{2} + \frac{p_y - p_x}{2} \cos 2\theta + q \sin 2\theta.$$

By eliminating  $p'$  and solving for  $q'$ ,

$$q' = \frac{p_x - p_y}{2} \sin 2\theta + q \cos 2\theta.$$

For the maximum values of  $p'$  and  $q'$ , as the angle  $\theta$  is changed, the conditions are respectively

$$\frac{dp'}{d\theta} = 0 \quad \text{or} \quad \frac{dq'}{d\theta} = 0.$$

Putting  $\frac{dp'}{d\theta} = 0$  gives

$$2q \cos 2\theta - \frac{p_y - p_x}{2} \cdot 2 \sin 2\theta = 0.$$

From this,

$$\tan 2\theta = \frac{2q}{p_y - p_x}$$

and

$$p'_{\max} = \frac{p_x + p_y}{2} + \frac{1}{2} \sqrt{4q^2 + (p_x - p_y)^2}.$$

In a plane at right angles to that on which  $p'_{\max}$  is acting there is a minimum normal stress

$$p'_{\min} = \frac{p_x + p_y}{2} - \frac{1}{2} \sqrt{4q^2 + (p_x - p_y)^2}.$$

These maximum and minimum normal stresses are called the **principal stresses**. On the planes on which the principal stresses act the shear stresses are zero.

In the case of brittle or semi-ductile materials the principal stresses are the dangerous stresses, if they are tensions. They will cause a break if they reach a value equal to the tension strength of the material. When the principal stresses are compressions, the shears must be examined.

Putting  $\frac{dq'}{d\theta} = 0$  we get

$$\frac{p_x - p_y}{2} \cdot 2 \cos 2\theta - 2q \sin 2\theta = 0.$$

Hence  $\tan 2\theta = \frac{p_x - p_y}{2q}$ . Comparing this value of  $\tan 2\theta$  with the value above obtained from putting  $\frac{dp'}{d\theta} = 0$ , we see that each is the negative reciprocal of the other. From trigonometry this means that the two values of  $2\theta$  differ by 90 degrees or the values of  $\theta$  differ by 45 degrees. Hence the maximum and minimum shear stresses lie in planes at 45 degrees to the principal stresses or maximum and minimum normal stresses. The value of the shears is

$$q'_{\max} = \pm \frac{1}{2} \sqrt{4q^2 + (p_x - p_y)^2}.$$

For ductile materials, this resultant shear is the dangerous stress. To avoid engineering failure of the piece  $q'_{\max}$  must be kept well under the yield-point strength in shear. Also for brittle materials where the principal stresses are compressions the resultant maximum shear stress must be kept under the breaking strength of the material in shear.

Ordinarily in combined loadings the normal stresses from the systems of loading will be present in one direction only. Either  $p_x$  or  $p_y$ , Fig. 67, becomes zero. This simplifies the formulas for  $p'_{\max}$  and  $q'_{\max}$ . They become

$$p'_{\max} = \frac{p}{2} + \frac{1}{2} \sqrt{4q^2 + p^2}$$

and

$$q'_{\max} = \frac{1}{2} \sqrt{4q^2 + p^2}.$$

**173.** The above discussion has assumed that the failure of a material occurs when the stresses  $q'_{\max}$  or  $p'_{\max}$  reach certain values, regardless of the relations of  $q$  and  $p$  which produce those values. The different combinations of  $q$  and  $p$  which result in single values of  $q'_{\max}$  or  $p'_{\max}$  set up different deformations in the material. On first sight the hypothesis that failure occurs when the deformation reaches a certain value is just as valid as the hypothesis that the failure occurs when the stress reaches a certain value. The experimental work of Guest, Scoble, Hancock and others has now made it practically certain that the magnitude of the stress is the controlling factor, not the magnitude of the deformation.

For a long time it was considered that the determining factor was the magnitude of the deformation rather than that of the stress. An "equivalent" tension stress was computed which would cause the same elastic deformation as did the actual stress system. This equivalent stress has the value

$$p_{\text{equivalent}} = \frac{1-m}{2} \cdot p \pm \frac{1+m}{2} \cdot \sqrt{4q^2 + p^2}.$$

In this expression  $m$  is Poisson's ratio, the ratio of relative decrease of thickness to relative increase of length in elastic tension loading. The value of  $m$  for steel is about 0.28. Hence

$$p_{\text{equivalent}} = 0.36 p \pm 0.64 \sqrt{4q^2 + p^2} \text{ for steel.}$$

It was assumed that failure would occur when this "equivalent" tension stress reached the yield point value for ductile materials or the breaking value for brittle materials. This assumption is now disproved; but the formula is still in rather common use.

## CHAPTER IX

### SPECIAL TESTS: IMPACT AND MINOR TESTS; HARDNESS; AND FATIGUE

174. While tests of materials in tension, torsion, transverse or compression loading serve the major requirements of the determination of properties, there are some properties not determined by these tests. There is also a field for special tests which try out the material for its fitness to a special service or to undergo certain special processes in manufacture.

175. A test in **direct shear** may be made to determine the **breaking strength in shear**. The method of making such a test has been described in Section 101. It has been shown that the information of a direct shear test can be obtained also from a compression test, a torsion test or a tension test, choosing the one of these tests which will give a shear break of the material in question.

176. **Impact or shock testing** is so much affected by the details of design of testing machine and form of test piece that no shock test has been accepted as a universal standard. In one field a special form of impact test called a "**drop test**" has been standardized and adopted. This is used on railway car wheels and axles. The specification calls for the taking without rupture of a certain number of blows delivered by a definite weight falling from a certain height. The falling mass is usually 1000, 1500 or 2000 pounds; the height of fall is up to 20 feet. The dimensions and construction of the apparatus are definitely specified.

In Europe an impact test on a "notched bar" test piece has been rather widely used. The form of the notch is shown in Fig. 26D or 26E. The piece is struck a transverse blow at the notched section. The energy absorbed in breaking the piece is measured. This test has never been accepted in America.

177. The **welding test** is used on materials which in manufacture are given hand welding. The test consists in attempting a simple

scarf weld under standardized conditions. The welds are broken in various ways to show the success and efficiency of the welding. Welding tests are hardly necessary when the welding in practice is done by the electric, acetylene or thermit processes. Quite as important as the success of the weld itself is the effect of the high temperatures on the metal near to the weld. This adjacent metal may be overheated and so made much weaker than the weld itself. For further discussion of overheating and its effects, see Chapter XIII.

**178. Bending tests** are a "rough and ready" way of estimating the ductility of metals. The test piece is bent around a stud having a diameter comparable to the thickness of the piece. The test piece should be bent with a lever, not pounded. The angle of bend before the piece cracks on the outside measures the ductility. The test may also be made by bending the piece back and forth through a definite angle, such as 90 degrees, counting the number of bends as measuring the ductility. It may be required that the piece be capable of bending down on itself without cracking. Bending tests may be made with the cold metal or with the metal heated to various temperatures. The purpose of the bending test is to find out whether the metal is ductile enough to be put safely through certain manufacturing processes.

**179.** Another kind of ductility determination is the "**drift test.**" It is made by driving a tapered steel pin so as to enlarge a hole drilled in a plate. The measurement is the relative increase in diameter of the hole before the edges crack or tear. This test is applied to plate steel which is to be used in structures fastened by riveting. Rivet holes in adjacent plates are often forced into alignment by the use of a "drift." In high-grade boiler work the use of a drift is forbidden; holes must be lined up with a reamer. Use of the drift makes a weaker joint.

**180.** The **hammer test** is made by tapping the material with a light hammer. The sound emitted tells whether there are defects, such as invisible cracks, in the metal. The hammer test is useful in locating defects in finished structures.

**181.** The **hardening test** is made on boiler-plate or ship-plate steel to find out if the carbon content is too high. A specimen is heated to a bright red heat and immediately plunged into cold water. It



should then stand the same bending test as required of an unhardened specimen.

**182.** An **abrasion test** is commonly used to find the durability of paving brick or stone. The testing machine is an octagonal barrel of specified dimensions. Into the barrel are put a certain number of brick and of hard cast-iron cubes of one- or two-inch size. The barrel is then rotated at a definite speed for a certain number of turns. The measurement is the percentage of weight lost by the bricks.

Abrasion testing machines for metals are now being built. As yet such testing can only be relative. Blocks of the metal to be compared are held down with the same pressure on the face of a grinding wheel, which is covered with some abrasive, such as sand or carborundum. The measurement is the loss of weight, or better of volume, per unit area of metal exposed to the grinding wheel.

**183. Hardness**, as a property of materials, is almost impossible to define satisfactorily. It may have reference to the ability of one material to cut another. This is called **cutting hardness**. Whether one material cuts another or not depends not only on the materials but on the forms of their surfaces where they are brought into contact. Where hardness measures resistance to wear or to indentation of surfaces in sliding or rolling contact with each other we speak of **surface hardness**. Hardness is a composite property, probably depending on the density and tensile and shear strengths of the material. Machines designed to measure hardness also indirectly measure the strength.

Some exclusively scientific instruments for measuring hardness use a minute plow made of a diamond point, and measure either the force required to plow a scratch of definite depth or the depth of scratch when plowing with a definite force. The hardness is measured by the ratio of force to the depth of scratch. Such instruments have no application in practical engineering work. They are called **sclerometers**.

Two types of engineering hardness testers are likely to be met with by the engineer, the pressure and the impact types, represented respectively by the Brinell ball tester and the Shore "Scleroscope."

**184.** For the Brinell ball test a flat place about an inch in diameter is cleaned and smoothed on the surface of the material. This flat

place must not be nearer to a corner or an edge of the material than about  $\frac{3}{4}$  inch. A hard steel ball  $\frac{5}{8}$  inch or one centimeter in diameter is pressed into the material with a definite load. This load is high enough so that the ball makes a permanent impression. The hardness is computed as the ratio of the load used on the ball to the area of the impression made in the material. In commercial work the size of the ball and the load to be used are standardized. Tables are furnished with the hardness testing machines to read "hardness number" from either diameter or depth of the impression.

To displace the metal with the ball and make the permanent impression which is measured, the metal tested must have been stressed considerably beyond its yield point. It is found in practice that the ultimate strength of a metal can be computed fairly accurately from the ball test hardness number. This gives to the test an additional importance, because it can be used as a field test of the strength of metals, which can be applied at any time without appreciable injury to the material and without taking the material out of its place in a structure.

**185.** The Shore scleroscope drops a tiny diamond-pointed steel hammer onto a small clean flat surface of the material to be tested. The hammer is guided by a glass tube surrounding it. The height of rebound of the hammer is observed on a scale behind the glass tube. If the material were perfectly elastic the hammer would rebound to the height from which it fell (except for guiding friction and air friction on the hammer). If some of the energy of the hammer is used up in doing work of permanent deformation on the material tested, then the rebound of the hammer is less complete. The maximum stress intensity at the hammer point during contact with the material may be about 500,000 pounds per square inch. No ordinary material can stand this pressure at a point on its surface without beginning to flow away from that point. The softer the material the less is the rebound height of the hammer. Hence a uniformly divided scale for measuring the rebound heights may be used as an arbitrary scale of hardness measurement.

**186.** Since the deformation produced by the hammer is both absolutely and relatively small, the stress reached in the material corresponds approximately to the yield point strength of the material.

In this the scleroscope differs from the Brinell test. The Brinell ball produces large local deformation of the material, corresponding to a point well out toward the breaking strength on the stress-deformation curve. On brittle materials where the yield-point strength may be considered, for the circumstances of the hardness measurement, the same as the breaking strength, the two types of hardness testers will give parallel results. On ductile materials the ball tester gets nearer to the cutting hardness. Cutting hardness varies in proportion to the breaking strength in shear. Impact hardness testers will show a hardness more than proportionate to the cutting hardness on cold worked metals because the "physical" stress intensity of the yield point is raised by cold working while the corresponding ultimate strength is not much changed. It would seem from the magnitudes of the deformations produced by the two methods of testing that the scleroscope would be the better instrument for finding "surface hardness," while the ball test should be better for finding "cutting hardness."

The scleroscope is particularly fitted for checking the uniformity of heat treatment in places where the test used must not injure the surface of the metal. The impression made by the scleroscope hammer is almost negligible. An example is the testing of the hardness of gear teeth or of the surface of a crank shaft bearing. When one manufactured piece has been found satisfactory in service, its hardness may be measured and subsequent pieces may be required to show that hardness.

**187.** When a piece breaks from the repeated application of small loads instead of the single application of a large load the break is called a **fatigue failure**. Even the loadings applied in service when repeated often enough cause fatigue failure. Hence the laws of fatigue failure, connecting range of stress with the number of applications of stress to break the piece, are of extreme importance in engineering design. We wish to use such stresses in service that the piece will wear out or be naturally retired by changes in practice before fatigue failure can occur. The number of applications of service load during the life of a machine, while large, is much smaller than might be expected. For example, take a high-speed engine going 500 r.p.m., 24 hours a day, 300 days per year, for 20 years. The total number of

revolutions would be 4,320,000,000. This is above the average, which is usually below 1,000,000,000.

**188.** When a piece is broken in one application of load the stress-deformation curve is of the type of Figs. 20 and 21. The area underneath these curves, between the curve, the deformation axis and a perpendicular dropped from the break points to the deformation axis, represents the total work done in breaking the piece in one loading. This work is distributed throughout all of the material. The internal damage done is approximately uniform in amount throughout the

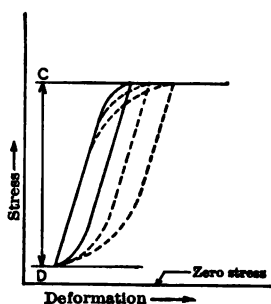


FIG. 68.

piece. The fatigue break is obtained by carrying the material repeatedly through a stress-deformation cycle such as is represented in *CGDHC* in Fig. 8. The area of the loop represents work done on the material per cycle. By repeating the cycle enough times the total work done upon the material, or the internal damage accumulated, may be enough to break the piece. The total accumulated work or internal damage from the cyclic loading, at the

break of the piece, may be more or less than the total work which would have been necessary to break the same piece in a single loading. In the cyclic loading the internal damage of the material is concentrated at certain weak spots instead of being uniformly distributed. On the other hand the material recovers or rests while loading is not being applied and also to some extent even during the cycle.

**189.** If a material is repeatedly carried through a stress range in cyclic loading such as *CD*, Figs. 68 and 69, the loop gradually becomes fatter. This is due to the progressive local breakdown of the material. The internal damage consists of shear slippage through the crystals. The weakest crystals are hurt first and by repetition of the loading form more and more slip planes, and become harder and more brittle by the working, until they give out completely. While the

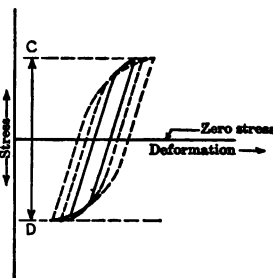


FIG. 69.

stress range of the material parts carrying load rise. Increase of the real stresses. This is the explanation

ing are "reversed," going the piece as a whole takes. During the progress of the stress cycle is the transverse of the loop takes place. The loop takes place as in. The loop may take a small perma-

usually that which would. Single application of the same metal gives a fatigue break. Then that a brittle material transverse loading.



B

be followed with a micro- of the fatigue failure of. Loaded loading with a stress. Fig. 70A shows the

appearance of the crystals after 5000 applications of the load. In certain crystals marked  $x$  in the figure, which happen to be weakest against the loading applied, slip bands have begun to appear. The other crystals are not yet affected. Fig. 70B shows the conditions after 40,000 applications of the loading. Slip bands have multiplied in the weakest crystals until they are broken down. From these points of initial failure, as from centers of infection, the breakdown of the material has spread until fracture of the piece is imminent.

**193. This study shows that the fatigue break goes through the crystals in such a way that on the fractured surface the original crystal structure of the metal would be made evident to the eye or to the microscope.** In this feature a fatigue break differs, if the material is ductile, from the fracture due to a single loading. In single loading to break of a ductile material the individual crystals are pulled out more or less to fibers so that the break cannot reveal to the eye the original crystal structure.

The difference between fatigue break and the break under a single loading, in the way that the original crystal structure is revealed, has caused a very persistent misconception of the reason for fatigue failures in service. It has been said that the cause of the service failure was "**crystallization,**" due to vibration or repeated stressing. The assumption is that the vibration caused the crystals to grow large and so weakened the piece. The truth of the matter is that in the case of the piece failing in service, when similar pieces stand up, the original crystal size of the piece which fails was larger. The crystals did not grow. The trouble originated in the manufacturing processes which left that particular piece "coarse grained." The circumstances of crystal growth will be discussed in Chapter XIII.

Vibration of parts of a structure sets up cyclic stress variation of high frequency and with possibly high stress range, the mass of the vibrating part acting as an inertia load on itself. If the frequency of the vibration, or the acceleration at the ends of the cycle, is known, the stress range can be computed from the mass and form of the part. Vibration stressing should be considered as one kind of repeated loading, and on this basis calculated and allowed for in design. Examples of such stressing are the connecting rods of engines, or axles of automobiles.

194. One factor in the commercial life of a machine part under repeated loading is the "resting" the metal gets when the machine is not running. It is well enough known in tension and torsion testing that a material recovers somewhat from the effects of a change of load when the piece is left standing at constant load. There is some parallelism between this resting of the metal after work and the resting of muscles after fatigue. Not only does the metal rest up from fatigue while the machine is not running; the rest effect enters even into the action of the cycle of loading and unloading. On account of the rest effect it may be expected that machine parts will stand a greater number of applications of their loading when allowed periods of rest than when worked continuously.

195. The main factor determining the life of a piece under repeated loading is the stress range in the cycle. The damage done per cycle, represented by the area of the stress-deformation loop (Fig. 68 or 69), changes rapidly with change of range of stress. This is illustrated in Fig. 71, for the case of soft steel. (Optical apparatus was used to

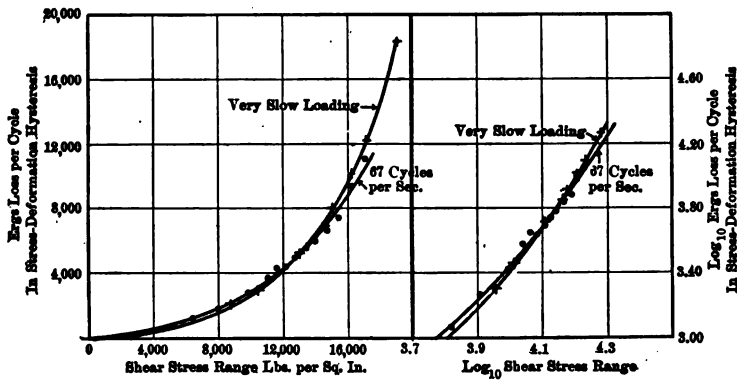


FIG. 71. Curves from Data of F. E. Rowett, "Elastic Hysteresis of Steel," *Proc. Roy. Soc.*, A 89, 1914, p. 528.

get the stress-deformation loop, and the area of the loop was measured.) An equation for the relation represented would be very hard to make. At low stresses the damage per cycle seems to vary approximately with the square of the stress range. As the stress increases the law changes; the exponent increases from two to three and four.

196. Fig. 71 indicates also that the damage per cycle depends upon the speed or the number of cycles per minute. There are two factors in the effect of speed. The shear slippage in the crystals, which constitutes the cyclicly reversed deformation, or the width of the loop, requires a certain time for its completion. Hence in rapid loading the width of the loop should be less with the same stress range. This in turn would argue longer life at higher speeds at a given stress range. The second factor is the rest effect within the cycle. This also depends upon time. At low speeds the material may rest appreciably while the stress is approximately constant at the ends of the stress range. The two factors oppose each other. For each material, and for each particular form of stress variation as a function of time, there should be a characteristic interaction of the two factors.

197. The curve of Fig. 71 has a form which might not be expected. It goes through the origin of stress and deformation. **This indicates that there is no stress range other than zero for which the damage per cycle is zero and the life of the piece infinite.** There is really no such thing as pure elastic action. If pure elastic action were possible, the zero value of damage per cycle should occur at the elastic limit stress. In other words, the curve would have a positive intercept on the stress axis. It has long been held, and still is a general opinion, that there is an "elastic limit" for a material such that any stress range less than that limiting value will give an infinite life in repeated loading. This view comes from thinking of the material as being internally structureless, homogeneous and isotropic. The material is actually an aggregate of great numbers of crystals often of different kinds and always of different orientations. During its manufacture to finished shape the piece has gone through processes which leave it with more or less of initial internal stresses from part to part and from crystal to crystal. Thought of in this way it is not so surprising that any stress, however small, from external sources, can find some crystals ready to be carried beyond their elastic limits and suffer damage. The individual crystals may have their ranges of pure elastic action, but the material which is an aggregate of the crystals has no such range.

198. The curve representing the relation between range of stress and number of cycles to break is the "endurance curve" of the ma-



terial. The endurance curve was first plotted with stress range as ordinates and number of cycles as abscissas as in Fig. 72 (full-lined curve). This yields a hyperbolic curve with a steeply descending branch near the axis of stress and a long flat branch approaching

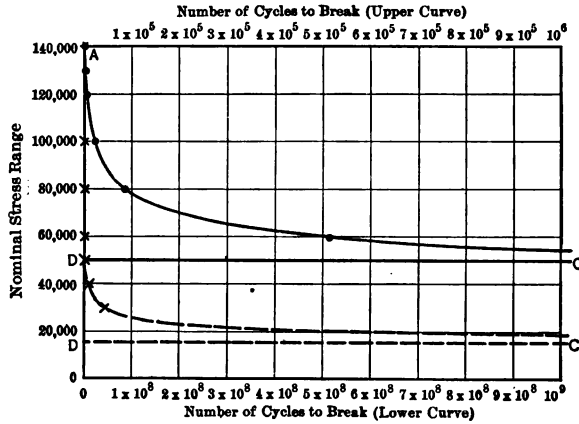


FIG. 72.

constant stress. This method of plotting the endurance curve gives an apparent possibility of drawing a horizontal tangent  $CD$  to the curve such that a stress range lower than  $CD$  would give infinite life to the material.  $D$  might then be called an elastic limit for perfect elastic action. The fallacy of this construction is seen when the same data is plotted on a different scale. It will then be found that the placing of the point  $D$  will depend on the scale of the drawing. (See Fig. 72, dashed curve, which plots the same data as the full-lined curve, with a change of abscissa scaling of 1000 : 1.) This method of drawing the endurance curve is valueless because it cannot represent on a single drawing the number of cycles at high stress ranges such as are used in tests, and at low stress ranges such as are used in practice.

What is wanted for the general plot is a curve that will show the ratio of the increase in life to the decrease in stress over the entire range from number of cycles equals one to number of cycles approaching  $10^{10}$  ( $10^{10}$  is probably greater than the number of cycles ever required of a piece of material during the commercial life of a machine). The desired plot is obtained by using log (stress range) as ordinate

vs.  $\log$  (number of cycles to break) as abscissa. Such a plot is shown in Fig. 73. On this logarithmic plot the endurance curve is nearly a straight line. The portion  $A_0 B_0$  is somewhat curved; the rest probably straight. The point  $B$  lies at a stress range practically twice the yield-point stress of the material.

If the material is given rest periods alternating with loading periods, the endurance curve shifts in the direction  $A_1 B_1 C_1$  and at the same

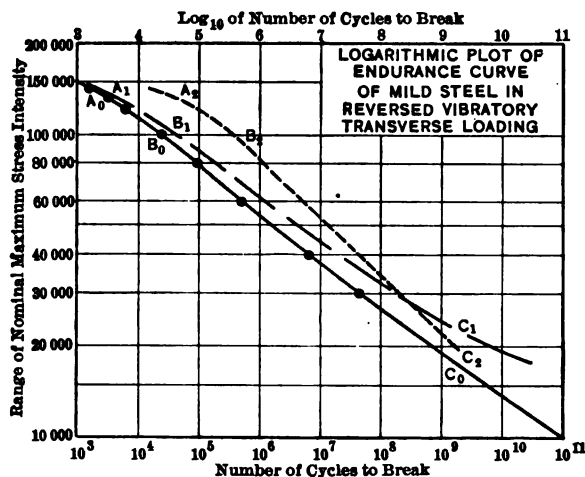


FIG. 73.

time becomes curved. The amount of the shift will be characteristic of the material and of the proportion of time allowed for rest.

High speeds change the endurance curve in the direction of  $A_2 B_2 C_2$ . The slope of the curve is changed as well as its position. At high stresses the endurance at high speeds seems greater than at low speeds. It is quite possible that the endurance curve for high speeds may cross the endurance for low speeds so that at low enough stresses the number of cycles to break will be less for high speed of loading than for low speed of loading. The curves of Fig. 71 indicate this. The amount of the speed effect is characteristic for each material.

199. Endurance curves for different materials differ in position and in slope when plotted as in Fig. 73. For example, the curve for copper is steeper than the curve for wrought iron. At high stress ranges, copper has as long a life as wrought iron. At low stress

ranges wrought iron is much superior. This knowledge must influence the methods of testing materials for fatigue resistance. It is not enough to find the relative life of two materials at one stress range, for the superiority of one over the other may be decidedly altered at other stresses. Comparison of materials for fatigue resistance must always be based upon portions of the endurance curves long enough to make a fair estimate as to the relations of the two materials at low stresses such as occur in practice.

200. A third method of plotting the endurance curve is to use stress range as ordinate against the reciprocal of the number of cycles to break as abscissa. This is shown in Fig. 74, curve A. Stromeyer has recently shown\* that if instead of the reciprocal of the number of cycles we use the fourth root of that quantity the major portion of the endurance curves becomes a straight line. This is shown in Fig. 74, curve B. The straight line has an intercept on the axis of stress

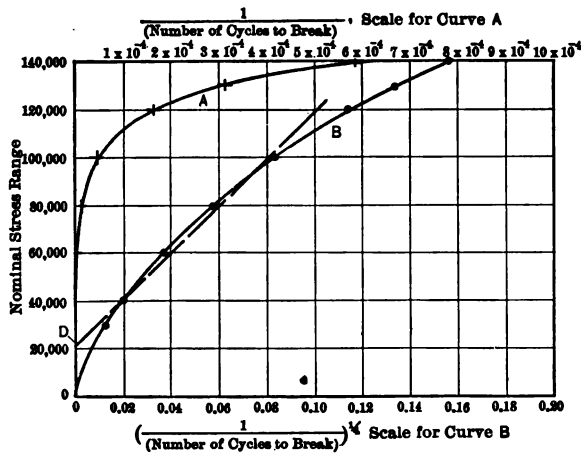


FIG. 74.

range marking an apparent limit of perfect elasticity for the material. The stress at this limit and the slope of the straight line give an empirical equation of the endurance curve. Stromeyer has determined these constants for a great number of engineering materials, making probably the first large collection of comparable data on fatigue

\* *Engineering* (London), June 1914, and *Proc. Roy. Soc.*

failure. For the purposes of engineering design the method used by Stromeyer is probably the best now available. The intercept  $D$ , Fig. 74, changes like the  $D$  of Fig. 72, with change of scaling of the curve.

201. It has long been the statement of empirical engineering that the proper material for withstanding fatigue is wrought iron or mild

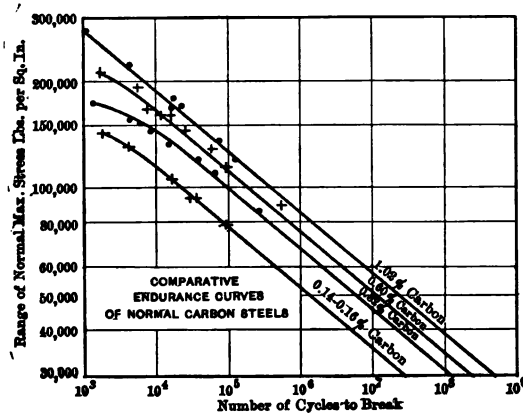


FIG. 75.

steel. Neither history nor actual experiment bear this out. The trend of specifications for railroad rails or automobile engine crank shafts, as examples of materials where fatigue must be guarded against, has been continually in the direction of higher carbon steels. The results of experiment are shown in Fig. 75,\* which brings out that the life at a given stress increases with the carbon content of the steel.

202. While the first studies of fatigue failure were made over fifty years ago, the importance of the subject has just begun to be realized by engineers. The amount of data available is as yet scanty and our knowledge of the subject only beginning.

\* From M.M.E. thesis on "Fatigue of Steel" by W. R. Wigley, Cornell University, 1915.

## CHAPTER X

### AGEING OF MATERIALS: CORROSION, RUST, WEATHER- ING, ROT, AND PROTECTION AGAINST THEIR EFFECTS

**203.** In the ordinary use of materials in structures it is customary to assume that the properties of the material are fixed and stable, unchanging with time. For most purposes this assumption is justifiable. It is not the exact truth; indeed the truth is that the properties do change with time. We should know the probable amount of this change and allow for it where necessary in construction.

**204.** Steels have two causes of slow change at ordinary temperatures. The chemical relationship of iron sulphide,  $\text{FeS}$ , to iron, changes at  $130^{\circ}\text{C}$ . Iron carbide,  $\text{Fe}_3\text{C}$ , and iron phosphide,  $\text{Fe}_3\text{P}$ , have allotropic changes in themselves indicated by changes in their magnetic properties at about  $200^{\circ}$  and  $400^{\circ}\text{C}$ . respectively. During the cooling of the steel at the end of its manufacture these various chemical changes are not completed. They go on very slowly during the life of the steel. Some steels become weaker and more brittle with age; others seem not to change. The steels may become brittle against shock while their static ductility does not change. No definite correlations between these chemical changes and the resulting physical changes have been worked out and we cannot predict whether a given steel will change or not. No allowance is made in design for these changes except that the factor of safety is supposed to cover them.

Heat-treated steels change slowly at room temperature. The rate of the change is negligible since apparently it is a matter of centuries for hardened steel to soften at  $20^{\circ}\text{C}$ . Hardened steels soften at a measurable rate at  $100^{\circ}\text{C}$ . and they soften rapidly at temperatures above  $200^{\circ}\text{C}$ .

On the other hand, Egyptian excavations have yielded tools of copper alloys which at the age of 4000 years show the same internal

structure as recently manufactured implements of the same alloys. In this case it seems that thousands of years have brought no perceptible changes.

Certain aluminum-zinc alloys as cast are in an unstable condition. Within a few months or years at room temperature they undergo a change accompanied by an alteration of volume which cracks them to pieces.

Tin is due for an allotropic change at  $18^{\circ}\text{C}$ . The form stable above  $18^{\circ}\text{C}$ . is the ordinary metallic "white" tin. The form stable below  $18^{\circ}\text{C}$ . is the powdery "gray" tin. The change of the white tin to the gray tin is erratic in starting. It requires either very low temperatures, about  $-40^{\circ}\text{C}$ ., or "seeding" with gray tin at temperatures nearer to  $18^{\circ}$ . Under ordinary climatic conditions a "tin" roof will remain in the white form. The change of white tin to gray tin is called "tin disease."

**205.** All structures put up with cement, especially with hydraulic cements, change properties with time over long periods. Concrete increases rapidly in strength for the first months after construction. Its strength may then drop although not usually to a dangerously low value. A subsequent rise of strength may extend over a period of about fifty years.

**206.** During "seasoning" or drying out of the sap moisture, woods generally gain in strength. Excessive drying may result in brittleness and loss of strength. The changes in moisture are accompanied by changes in volume which cause "checking" and similar breaking up of the structure. When the wood is subject to shear along the grain these changes may be serious. The volume change or "shrinkage" of wood takes place in directions perpendicular to the grain rather than along the grain.

**207.** We have been dealing above with changes of internal origin in the material. Changes of external origin are often more important. Against these changes protection can often be provided.

**208. Metals are affected by corrosion or rust.** The chemically pure metals are little affected by acids. If pieces of two metals are put into an acid and the two pieces are connected externally by a conductor, an electric battery is made and rapid "electrolytic" corrosion of one metal results. The commercial metals are not pure.

They consist of aggregates of crystals of different kinds, and even within the individual crystals the composition may be variable. If such a metal is put in contact with a liquid containing acid, even if there is no more than a film of moisture, local electrolytic corrosion takes place from one crystal to another or from one part to another of the same crystal. If the products of corrosion are continually removed, the metal may be ultimately destroyed. Some of the metals, like aluminum, copper or lead, form adherent coats of oxide which stop further corrosive action. Corrosion and rusting are so much affected by the nature of the material surrounding the metal that it is impossible to give specific values for the resistance of metals to corrosion.

**209.** The most important corrosion of metals is the rusting of iron and steel. This takes place whenever moisture in contact with the metal becomes electrolytic either by acids or alkalis. The chemical changes involved are complex and variable. The ultimate product is a more or less hydrated red oxide of iron. Rust destroys iron and steel at such a rate that protection is always necessary. The oiling of machinery is a protection. Other protections are galvanizing, which is coating with zinc, tin plating and painting. The black iron oxide,  $\text{Fe}_3\text{O}_4$ , formed on the surface of steel or iron during hot rolling, acts as a protective coating while it adheres. When it is broken it hastens the rusting. Zinc coating of iron is a very good protection because electrolytic action between zinc and iron is at the expense of the zinc. Electrolytic action between tin and iron is at the expense of the iron. Hence a good tin plate must be heavy and thoroughly adherent to the iron.

**210. Weathering of stone and concrete** is due principally to action of water. The materials are porous, and absorb water which may then act upon them chemically or physically. A physical action of importance is the expansion as the water is frozen to ice. This expansion is strong enough to break off the surface of the stone or concrete. Steel bedded in concrete may be rusted by the water. Concrete resists weathering (or the action of sea water) in proportion to its freedom from pores through which water can enter. The requirement of maximum density in a concrete, which is specified in order to get high strength, is also a requirement which would lead in the direction of freedom from pores, and of resistance to weathering.

**211. Woods are subject to wet and dry rot, growths of fungi and attacks by insects.** The durability of wood depends largely upon initial seasoning. The drying out of the sap is essential to the preventing of dry rot, which is due to fungous growths that in their beginning feed upon sap. A partial prevention of rot and insect attack may be obtained by proper seasoning of the wood. Complete prevention requires in addition to proper seasoning the impregnation of the wood throughout its volume with some kind of chemical preservative. Creosote may be cited as an example of such a preservative. Painting prevents the access of moisture and germs or spores to the outside of the wood and so protects against wet rot, fungous attacks and insects.

**212. Paints are used as protective coatings both on metals and wood.** The paint should form an adherent coating which should be water-tight, preventing the access of moisture to the metal or the wood. The part of a paint which forms the impervious coating is called the vehicle or binder. Color is given to the paint, for the sake of appearance, and the binder is made to go further, by the use of a base or pigment. Common bases are white lead, zinc white, red lead, iron oxide and graphite.

These bases may have their colors modified by the addition of various pigments. The most common vehicle is linseed oil. This oil slowly oxidizes in the air, making a tough and water-tight film surrounding the particles of base or pigment and adhering to the metal or wood. This oxidation of the oil is called the "drying" of the paint. The drying may be hastened by "driers." Staining of wood may be done with pigments carried in volatile "solvents" such as turpentine.

Metals may be given a protective coating of asphaltum. The asphaltum for this purpose must be naturally stiff when cold and little softened by heat up to the boiling point of water. Cast iron and steel water pipes are coated by dipping in the hot molten asphaltum. Bridgework may be painted with the hot asphaltum or with a cold solution of asphaltum in a volatile solvent.



## CHAPTER XI

### CHOICE OF MATERIALS, WORKING STRESSES AND FACTORS OF SAFETY; SELECTION OF TESTS; SPECIFICATIONS

**213.** The choice of a material for a given construction is determined ultimately by financial considerations. We use the cheapest material which will **satisfactorily perform** the service demanded. As a result of the relation of their properties, which determine possible uses, to their prices, which determine the selection, we have come by experience to know for what uses each material is fitted. We would not, for instance, make a pier for a bridge out of solid cast iron because we know that we can get the necessary strength of structure from stone masonry, or concrete, or skeleton steel work, much more cheaply. In the building of a lathe certain sizes of parts are necessary to give rigidity to hold the cutting tool accurately to place while it is working. With this size of parts steel would make a stronger structure than is necessary. Cast iron will have ample strength and is a cheaper material.

Special properties indicate special uses. Two examples will be discussed, cast iron and steel. Cast iron is the cheapest of the metals. It is readily shaped by casting, which is the cheapest of shaping processes. It is moderately easy to machine for finishing. The porosity of the finished surfaces, due to falling out of graphite crystals, makes cast iron hold an oil film for lubrication better than do the other metals. Cast iron is of rather low strength and is brittle. This combination of advantages (and disadvantages) gives cast iron very wide use in construction of machinery, where weight of parts is desirable rather than otherwise. With heavy parts low strength and brittleness of material are not such serious disadvantages.

Next in cheapness to cast iron are steels. Steels have the advantage over cast iron of much higher strength and much greater ductility. The shaping of steels by casting, forging, rolling or machining is

more expensive than that of cast iron. Steels are used for general structural work where weight of structure has to be considered. Wherever weight of the material becomes a load on the material, high strength per unit weight is essential.

**214. Factors of safety measure the margin between ordinary working conditions and those loads or shocks which would be able to cause failure of a structure.** The factor of safety changes through wide ranges with change in nature of the material and with change in the loading conditions. By convention a factor of safety is stated as the ratio between the engineering maximum strength of the material in the kind of loading employed and the working stresses for which the structure is designed. The following table gives typical factors of safety.

Material	Factor of safety			
	Dead loads	Live loads		
		Temporary structures	Permanent structures	Shock loads
Iron and steel (ductile metals).....	4	4	4 to 6	10
Cast iron (brittle metals).....	4	4	6	10
Wood.....	4 to 5	4	10	20
Brick work (cut stone).....	4	.....	10	.....
Masonry.....	5	.....	15	.....

**215.** The considerations entering into the choice of a factor of safety or the working stresses assumed in design may be analyzed to show why the factor of safety has such various values.

(1) **Our knowledge of the material and the properties it will have is subject to uncertainty.** This uncertainty increases in going from top to bottom of the table above. The factor representing its effect on factor of safety varies from 1 to 2 or 3. **There are uncertainties in the qualities of workmanship in construction,** and the consequent initial stresses in the structure. **There are also uncertainties as to changes in the service loadings and unforeseen contingencies.** The Brooklyn Bridge is an example.

(2) The effect of corrosion or change of the material with age

requires larger factors of safety for permanent than for temporary structures.

(3) **The method of computing the stresses changes the factor.** From dead loads we compute the true stresses that are probable. Live loads applied without shock (for example a rolling load crossing a bridge at low speed) actually set up stresses twice as great as a dead load of the same weight. Instead of recognizing this directly in our computations of stresses we calculate the stresses as if the load were dead and increase our factor of safety in the ratio of  $1\frac{1}{2}$  or 2 to 1. This change of factor of safety for live loading must also increase toward the bottom of the table because of the increasing chance that the material will disintegrate from vibration. Just as live loads from given weights cause twice as much actual stress as dead loads of the same weight, shock loads cause still larger stresses. The velocity or energy of the shock should be considered. The higher the velocity the greater should the factor of safety become, if the computation does not recognize the increased stresses.

**Shock loading should be figured from the energy which the material is required to absorb per unit volume.** This energy-absorbing power of the material is measured by the area underneath the engineering tension test curve of the material. For steels the energy absorbed in permanent deformation beyond the yield point is from 50 to 100 times as great as the energy absorbed in elastic loading at ordinary working stresses. For brittle materials, with engineering tension test curves nearly straight lines to the break point, the same factor of safety would require that the ratio of breaking strength to working stress should be of the order of magnitude  $\sqrt{50}$  or  $\sqrt{100}$  to 1. When the relation between stress and deformation follows a straight-line law the energy absorbed in loading to a given stress varies with the square of the stress. If, therefore, the energy absorbed at ordinary working stresses is to be  $\frac{1}{50}$  or  $\frac{1}{100}$  of that required to break the material, the ratio of breaking stress to working stress must be  $\sqrt{50}$  or  $\sqrt{100}$  to 1.

(4) **If the piece is subjected to repeated loading or variation of loading, the chance of fatigue failure must be considered.** To cover this the factor of safety appropriate for dead loading must be increased by multiplication by a factor taking account of the range of stress in

the variable loading, the number of times the variation is likely to be repeated, and the endurance properties of the material. The range of stress must be computed as accurately as possible in true values, such careless assumptions as those made in calculating live-load stresses being ruled out. The mathematical form of this factor may be suggested to be

$$\left(1 + \frac{\text{range of stress}}{\text{engineering elastic limit stress of material}}\right)^n,$$

in which  $n$  may for a first approximation be taken as unity, when the load is to be repeated a great many times. The maximum numerical value of allowable stress would be reduced from that for dead loading by division by the fatigue factor above. When in applying this formula the computed allowable maximum numerical stress comes out greater than one-half the stress range, the absurdity is explained by the knowledge that the stress range assumed would cause fatigue break before the structure had fulfilled its proper life and service. The exponent  $n$  of the formula varies between zero and 1, approaching 1 as a limit as the number of probable cycles of loading approaches infinity, and approaching zero as a limit as the number of probable cycles decreases toward 1.

For vibration as a cause of fatigue failure see the last paragraph of Section 193.

(5) The working stresses must always be well within the yield point or engineering elastic limit of the material, for appreciable permanent deformation of the material must not take place. With carbon steels not heat treated the ratio of yield-point strength to engineering maximum tension strength is about 0.55, nearly independent of the carbon content of the steel.

The inverse of this ratio, approximately 2, must enter into all factors of safety. The ratio of engineering elastic limit to breaking strength of cast irons is about 0.35 to 0.50, the inverse from 3 to 2. With brasses and bronzes and heat-treated steels the ratio of yield point to engineering maximum tension strength may be anywhere from 0.3 to 0.8 depending on the material and its preparation. The reciprocal numbers, 3.5 to 1.25, enter as components into the factor of safety for the material. The fact that this ratio is approximately

two for unheat-treated steels has led to a common, and unjustified, assumption that the ratio is two for all materials.

The components which multiplied together give the factor of safety may now be summarized.

(1a) Knowledge of material and workmanship.

Material	Best workmanship	Ordinary workmanship
Steels	1	1.5
Cast iron	1.5	2
Woods }	2	4
Masonry }		

(1b) Contingencies.

1 to 5, depending on the nature of the structure. Low for ductile materials, high for brittle.

(2) Ageing.

1 to 1.25 for metals.

2 to 5 for woods.

(3) Method of stress computation.

1 for loads all "dead" to 2 for loads all "live"; for shock loads as high as 5.

(4) Repeated loading or fatigue.

1 for load all "dead," no variation of stress, to 1.5 for stress all variable between equal negative and positive limits.

(5) Ratio of engineering maximum strength to engineering elastic limit or yield-point strength.

1.25 to 3.5, depending on the material.

**The last and most important factor is judgment and experience.**

216. Purchase of materials is usually to specifications, the meeting of which is checked by inspection and tests. On contracts of importance the purchaser's inspector is present at the factory producing the material. A second inspection occurs on receipt of the goods. Testing may be done by the maker, by the purchaser or by third parties according to the agreement. Methods of inspection and methods of tests are now very generally standardized. These stand-

ard methods should be adhered to closely. Specifications should be so drawn that there is no chance for "faking" the tests. For instance the ductility measurement of a tension test changes with the shape of test piece; therefore the shape of the test piece must be specified.

The tests required should be such as to bring out not only the general properties of the material but also any special properties which may be needed in the use of the material. As an example of this selection of tests, the following table shows the recommendations of the American Society of Mechanical Engineers.

TABLE SHOWING TESTS REQUIRED

(Required test denoted by X)

Material used for	Tension	Compression	Transverse	Torsion	Impact	Welding	Bending	Hardening	Forging	Abrasion	Punching
Railroad rails.....	X				X		X				
Railroad car-axles.....	X				X		X				
Railroad car tires.....	X				X						
Shafting.....	X		X	X							
Building, wrought iron.....	X	X	X			X	X				
Building, low steel.....	X	X	X				X	X			
Building, high steel.....	X	X	X				X	X			
Boiler, wrought iron.....											
Boiler plates.....	X		X				X		X		X
Boiler, shape-iron.....	X					X	X		X		X
Boiler, rivet-rods.....	X						X		X		
Boiler, low steels.....	X						X	X	X		
Ship materials.....											
Ship plates.....	X						X				
Ship rivets.....	X						X		X		
Wire.....	X						X*				
Wire rope.....	X				X†						
Cast iron.....	X	X					X				
Copper and soft metals.....	X	X					X				
Woods.....	X	X	X								
Stones.....		X	X								

\* Repeat in both directions — also by winding.

† Longitudinal.

The following is a brief list of the different authorities which have brought out special specifications for materials and testing:

American Steel Manufacturers. — Specifications for structural steel, rivet steel, bridge steel and structural cast iron.

American Boiler Manufacturers Association. — Specifications for cast iron and for all kinds of steel used in boiler construction.

“Lloyd’s.” — Specifications for ship steel.

American Foundrymen’s Association. — Specifications for gray cast iron.

American Water Works Association. — Specifications for cast-iron water pipe.

American Society of Automobile Engineers. — Specifications for practically all materials used in automobile construction.

American Society for Testing Materials. — This society being the official body dealing with the subject of materials as a whole, has collected and systematized specifications for practically all materials in engineering use. Any one desiring to know what may be proper standards and specifications for any material, so far as standards have been adopted, may communicate with this society.

## CHAPTER XII

### THE NATURE AND ORIGIN OF THE STRUCTURE OF ALLOYS

**217.** The metals used in engineering are alloys. The term alloy is used to distinguish a metal to which chemical elements other than the principal one have been intentionally added. The addition takes place in the liquid state. The reason for using alloys rather than pure metals is that the engineering properties are almost always better in the alloys. Metals as near to chemical purity as possible are used for electrical conductors. Examples are copper, aluminum and iron. Iron nearly pure also finds some uses because it rusts or corrodes much less rapidly than the alloy of iron and carbon called steel. The strength of alloys is almost always better and in some cases far better than that of the pure metals.

The composition of an alloy means the percentage by weight of the different chemical elements present. These chemical elements are combined with each other in various ways. These combinations build crystals. The aggregation of the crystals in their various kinds, sizes and relations to each other constitutes the structure of the alloy. For the understanding of the properties of the alloy one must know the structure and the percentages by volume of the different kinds of crystals in the structure.

**218.** The study of the origin and nature of the structure of alloys goes back to the chemical unit of matter, the atom of the chemical elements. The chemical atom is a body of extreme minuteness, its diameter being approximately  $10^{-8}$  cm., or 0.000,00025 in. An isolated atom may be considered as surrounded by three fields of force, as indicated in Fig. 76. The general arrangement of these fields of force around the atom is characteristic of the atom. The innermost field with lines of force in closed circuits is an electrostatic field of high intensity. Beyond this is a weaker field of electromagnetic action. Thirdly, there is a gravitational field with lines of force running straight out from the atom. When atoms are at considerable dis-



tances from each other the only force attracting them to each other is the gravitational force, because the lines of force in the other fields are closed circuits. If two or more atoms are brought very close together and if at the same time their electrostatic fields happen to match in position and direction, then a **chemical compound** or **molecule** is formed by the mating of the electrostatic fields. Within their radius of action these forces are very intense. If the electrostatic forces are not all used up in the formation of a chemical compound, the remaining forces may cause the simple molecules to combine with other molecules into complex molecules.

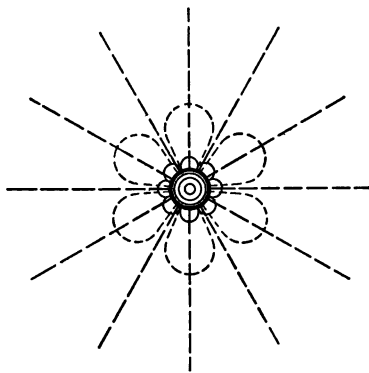


FIG. 76.

**219.** In the gaseous or vapor state of matter the average distance between molecules is large as compared with their own dimensions. The molecules are in rapid motion, with frequent "collisions" with one another. The orbits of the molecules with regard to each other are hyperbolic with the curvature at "collision" so sharp that the paths between collisions are practically straight lines. Except during collision the only forces acting between molecules are the gravitational forces. During collision the electromagnetic forces are acting and in some cases even the electrostatic.

**220.** As the gas or vapor cools off, the average velocity of the molecules decreases and the average path between collisions becomes more curved. At some stage in the cooling a point is reached where the average energy of motion of the molecules is not sufficient to take them out of range of the electromagnetic forces after collision. The orbit of the molecules with regard to each other suddenly changes from the hyperbolic to the elliptical type. This constitutes the change from the vapor to the liquid state of the material. The energy given out as the nature of the orbit and the controlling forces change is called the **latent heat of liquefaction or vaporization**. In the liquid state the molecules are continually in contact with one another and in motion with regard to each other and with regard to the body as a whole.

**221.** The electromagnetic field of the molecules has polarity. This polarity is complex. The directions and intensities of the forces around the molecules are characteristic of the molecule. There is a tendency for the molecules to build up on each other in chains or groups in definite patterns controlled by the directions of the more intense electromagnetic forces. When the temperature of the liquid is high the energy of motion of the molecules breaks up these groups as fast as they begin to form. As the liquid cools the energy of motion of the molecules decreases. A point is finally reached at which the tendency to grouping is too strong for the kinetic energy to break up the groups. Then at great numbers of points through the body of the liquid the molecules begin to build up on each other in geometric patterns determined by the polarity of their electromagnetic fields. These groups constitute **crystals**. The cessation of wandering motion of the molecules constitutes the change from the liquid to the solid state. In the solid state molecules are fixed in position with regard to the body of which they are a part. Their heat motion is a vibration about this fixed position. When the grouping of the molecules into crystals by the directed electromagnetic forces takes place, heat energy is set free, which is called the **latent heat of freezing or solidification**.

**222.** A liquid may change to a solid by the slowing down of the heat motion of the molecules without the grouping of the molecules into crystals. The solid so obtained is "**amorphous**." An amorphous solid has no latent heat of freezing or melting, because the character of the forces between the molecules has not been changed from that of the liquid state. Physically an amorphous solid may be considered as a liquid of very high viscosity (resistance to flow). Ordinary glass is an example of an amorphous solid.

**223.** The kinetic energy of the molecules, the intensity of which we feel as temperature, is not the same in all the molecules of the body. At any given moment all velocities from zero toward infinity are represented. There are always some molecules of a solid or liquid with sufficient energy to break away from the hold of the electromagnetic forces and form a vapor phase of the material. Vapor pressure increases with temperature because with rise of temperature a greater proportion of the molecules are able to escape from the

solid or liquid. Any state of a material, vapor, liquid or solid, may be considered to contain traces of the other states. Some phenomena of the solid state, for example, "diffusion," may perhaps be explained on this basis. The lowest energies of motion of the molecules belong to the solid state, intermediate values to the liquid state, and the highest values to the vapor state.

**224.** When crystals begin to form in the liquid there is a selective grouping of the molecules. To build into a definite geometric pattern it is necessary that their size, and the direction of their polarities, should be alike from one to another. The size and the direction of their polarities vary from one kind of molecule to another and are characteristic of the molecule. Hence crystals as they form tend to be built of chemically similar molecules, because dissimilar molecules will not fit the patterns. If it happens that in the liquid there are two kinds of chemical molecules closely alike in size, and in direction of their polarities, it is possible that these two kinds of molecules will build up together in the single crystals, since the pattern may be common to both. The two are then said to be in **solid solution** with each other. Among the metals, solid solutions are extremely common.

**225.** The word crystal ordinarily conveys an idea of a portion of material having the form of some geometric solid, bounded by plane faces. The crystals of the metals rarely have such external forms. As crystallization proceeds from the liquid, the crystals grow outward from their nuclei by organization of molecules from the adjacent liquid. Crystal growth ends by exhaustion of the supply of the molecules which will fit into the pattern of a specific crystal, and finally by exhaustion of the liquid. The external forms of the crystals are then due generally to their interferences in growth with each other. The external shapes may be called accidental and have little to do with the internal geometric structure of the crystals.

**226.** Many of the chemical elements and many chemical compounds have more than one possible crystal form. These different forms are called **allotropic** forms of the substance. Allotropy means in general the power of a substance recognized as chemically the same to appear in different physical forms. Allotropy may be due to: (1) polymerization, or change of the number of atoms in the molecule by doubling, tripling, etc.; (2) rearrangement of the atoms in the molecule without

change in number; (3) both of these actions. Rearrangement of atoms in the molecule may cause so large a chemical change that we consider the substance to be itself changed. In the case of oxygen and ozone, or the various forms of sulphur or of phosphorus, allotropy is known to be due to change of number of atoms in the molecule.

The possible diversity of physical properties between allotropic forms of the same substance is well illustrated in the case of carbon. Diamond has a density of 3.52, and is one of the strongest and hardest substances known. Diamond dust is one of the best of abrasives. Graphite is just as much carbon as diamond is. The density of graphite is 2.25. It is so weak and soft that it is commonly used as a lubricant.

Allotropy is common in the engineering metals. Copper, aluminum, magnesium and lead each have one form. Manganese and antimony probably have two forms. Tin has two or three forms. Iron, cobalt, nickel and zinc each have three forms. Concerning chromium, vanadium, titanium and tungsten our information is not definite; we recognize one form of each. The chemical compounds of the metallic elements with each other or with other chemical elements are also subject to allotropic change. Examples occurring in steel are the cases of  $\text{Fe}_3\text{C}$  and  $\text{Fe}_3\text{P}$ .

Change of allotropic form is accompanied by a heat change. In the crystalline condition the bonding of the molecules to each other is changed in direction or intensity, or both. In many ways these changes are similar to the phenomena of change of state. Change of allotropic form, like change of state, involves a latent heat, changes of density or specific volume, and in general a discontinuous alteration of the physical properties.

**227.** To understand the growth of the crystalline "structure" of an alloy during its freezing and the relation of structure to composition, we will study first the simple case of the series of alloys made by mixing lead and tin in various proportions. Pure lead freezes at  $326^\circ \text{C}$ . There being only one kind of molecule present in the liquid, crystallization is completed at the same temperature at which it begins. Below  $326$  degrees the lead is solid, and is an aggregate of crystals. Similarly pure tin freezes at  $232^\circ \text{C}$ . Let now some tin be dissolved in liquid lead giving the temperature and composition of point *A*, Fig. 77.

When this liquid alloy cools it does not begin to freeze at the freezing temperature of pure lead, but at an appreciably lower temperature. This is because the presence of the tin molecules in the liquid produces an alteration of the internal forces in the liquid, delaying the freezing out of the lead molecules to a lower temperature than for pure lead.

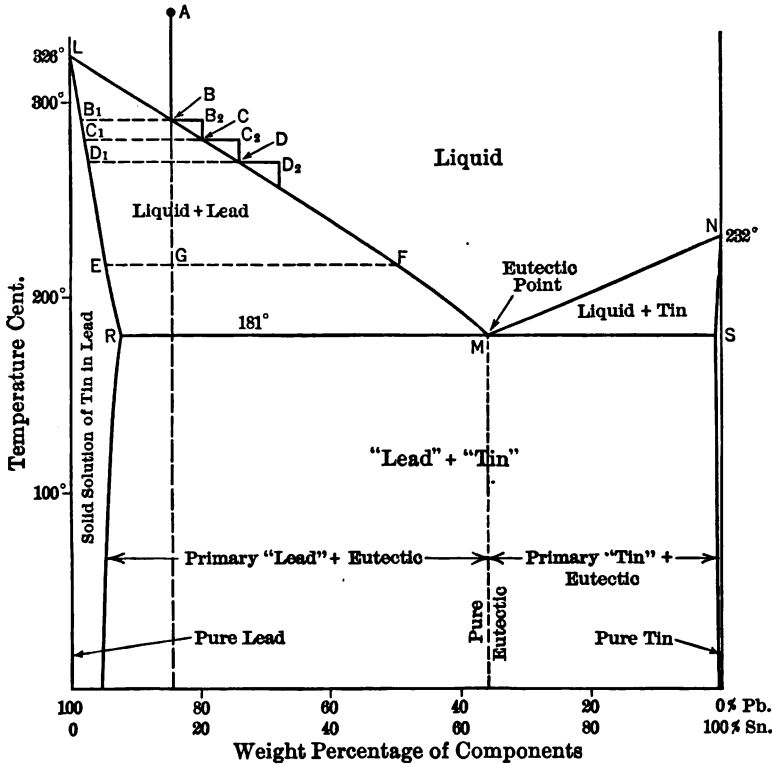


FIG. 77.

At the point *B* freezing begins. The material separating as crystalline solid is nearly pure lead, for lead does not take much tin into solid solution, while liquid lead has more or less unlimited solubility for tin. The composition of the solid crystals growing in the liquid when freezing begins at the point *B*, Fig. 77, is represented by the point *B*<sub>1</sub>, which represents lead with a little tin in solid solution. *B*<sub>1</sub> is at the same temperature as *B* but has the composition of nearly pure lead. This abstraction of lead, while leaving the tin in the liquid,

makes the liquid richer in tin and shifts the composition of the liquid to the point  $B_2$ . In turn this new liquid cools to the point  $C$  and freezes out lead (with a little tin in solid solution) of the composition and temperature  $C_1$ , while the remaining liquid shifts to  $C_2$ . Actually this series of processes goes on in infinitesimal steps. As the mass cools molecules of lead continually join the primary lead crystals, increasing them in size and number. The liquid may be said to be continually feeding lead to the primary lead crystals. The composition and temperature of the liquid remaining travels downward along the line  $LM$ . At any temperature, such as that of  $EGF$ , the material which started as homogeneous liquid at  $A$  will have split into two parts, crystals of lead (with a little tin in solid solution) of the composition  $E$ , floating in a remaining liquid of composition  $F$ . The average composition remains on the vertical line  $AG$ . The line  $LBDFM$  is a "liquidus" line above which the alloy is entirely liquid and below which the alloy is partly or completely solid.

Starting from the melting point of pure tin at  $N$ , Fig. 77, is a similar liquidus line  $NM$  representing the temperatures at which tin begins to freeze out of the liquid alloys. The addition of slight amounts of lead to tin lowers the temperature of beginning of freezing. Tin freezes out first leaving the lead in the remaining liquid. This action is quite contrary to the first expectation, which would probably be that addition of lead would raise the freezing point of tin and that the first material to freeze out would always be lead, since lead has a higher freezing point than tin.

228. At the point  $M$  the two liquidus lines cross. Each cuts off the other from extension to lower temperatures. A liquid solution reaching the point  $M$  in composition and temperature must deposit lead crystals because  $M$  lies on the line  $LM$  and tin crystals because  $M$  lies on  $NM$ . Hence the freezing must be complete at  $M$ . The intersection of two liquidus lines at such a point as  $M$  is called a "eutectic" point.

It is now possible to complete the description of the freezing of the alloy which started to cool from the point  $A$ . It cools with change only of temperature to the point  $B$  on the line  $LM$ . At  $B$  the freezing out of primary lead crystals begins. At temperatures below  $B$  the "melt" feeds lead to these crystals causing them to grow large

and numerous. The remaining liquid continually decreases in amount. It also changes in composition and temperature along the path *BCDFM*. Until *M* is reached only lead (with a little tin in solid solution) separates from the melt; the bulk of the tin remains in the liquid solution. At *M*, because of the crossing of the other branch *NM* of the liquidus curve, **simultaneous** separation of lead and tin crystals from the melt occurs. Since the two kinds of crystals are forming at the same time, but the molecules are unlike, the two kinds of crystals interfere with each other's growth. That portion of liquid which freezes at the eutectic point *M* therefore produces an intimate mechanical mixture of minute lead and tin crystals. This eutectic mixture of minute crystals of both lead and tin will surround the large primary lead crystals, which will appear to be bedded in a "matrix" of the eutectic structure. Below the temperature of the eutectic point the alloy will be completely solid.

**229.** The freezing-point diagram may now be laid out into fields showing the different phases through which the alloys go in passage from the liquid state to ordinary temperatures. Above *LMN*, Fig. 77, the alloys are entirely liquid. In the field *LRM* the alloys consist of a mixture of primary lead crystals and liquid. The composition of this liquid is determined by the liquidus line and the temperature. For example, — at the temperature *EGF* the liquid present has the composition *F* for any alloy in the range of compositions between *E* and *F*. Similarly the field *MNS* has the two phases, liquid and primary tin crystals. The composition of the liquid will again be determined by the liquidus line (in this case *NM*) and the temperature. Below *RS* the alloys are completely solid, consisting of mechanically mixed lead crystals and tin crystals.

**230.** The structure of the solid alloy is not completely described by calling it a mechanical mixture of lead and tin crystals. The peculiar eutectic structure comes at 37 per cent lead. On the one side of the eutectic the structure of the alloys consists of large primary lead crystals surrounded by eutectic mixture of small lead and tin crystals. On the other side of the eutectic the structure appears much the same, but the large primary crystals are of tin. The proportion of the large primary crystals, as composition changes through the series of alloys, decreases from both sides toward the eutectic.

At the eutectic composition there are no large primary crystals of either kind.

The changes in size and arrangement of the crystals are accompanied by important changes in physical properties. Fig. 78 shows

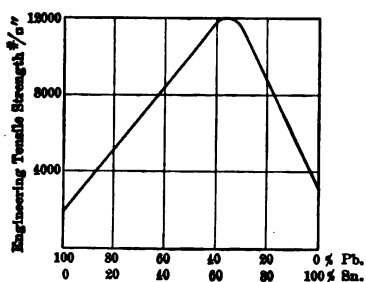


FIG. 78.

the engineering tensile strength of the lead-tin series of alloys. The importance of the eutectic structure, or more properly of fineness of crystal size, is shown by the high strength of the eutectic as compared with the pure metals.

231. When the freezing-point diagram is extended to represent changes occurring after freezing as well as during

freezing, it becomes the “**equilibrium**” diagram of an alloy series. The freezing-point diagram is really but the upper portion of an equilibrium diagram. The equilibrium diagram is a complete presentation of the various phases (forms and conditions) in which the components of an alloy appear at various compositions and temperatures. Theoretically the equilibrium diagram is concerned only with those phases which would be chemically and physically stable under the specific conditions of composition, temperature and pressure. In practice, however, we also often represent upon the equilibrium diagrams phases which are unstable. There are many cases in the engineering alloys where unstable forms or phases decompose so slowly that for practical purposes they may be considered as permanent.

232. The most important equilibrium diagram for engineers is that of the alloys of iron and carbon. Ordinary or “carbon” steels consist mostly of iron alloyed with small amounts of carbon and with amounts of other elements which are practically negligible and may for most purposes be considered as impurities. Cast irons are more complex alloys containing significant amounts of iron, carbon, silicon, phosphorus and sulphur. Even in cast irons the general nature of the material can be studied in terms of the iron and carbon alloy.

Fig. 79 shows approximately the iron-carbon equilibrium diagram as at present known. Some of the sloping lines of the diagram are drawn straight when they should be slightly curved. The vertical



and horizontal lines should be straight as they are drawn. The abscissa is the per cent by weight of carbon; it is to be understood that the remainder of the alloy is iron. Since the most important form in which carbon occurs in these alloys is the carbide of iron,  $\text{Fe}_3\text{C}$ , the top of the diagram is given an abscissa scaling in weight percentage of possible  $\text{Fe}_3\text{C}$ . When carbon is not present in the alloy as graphite,

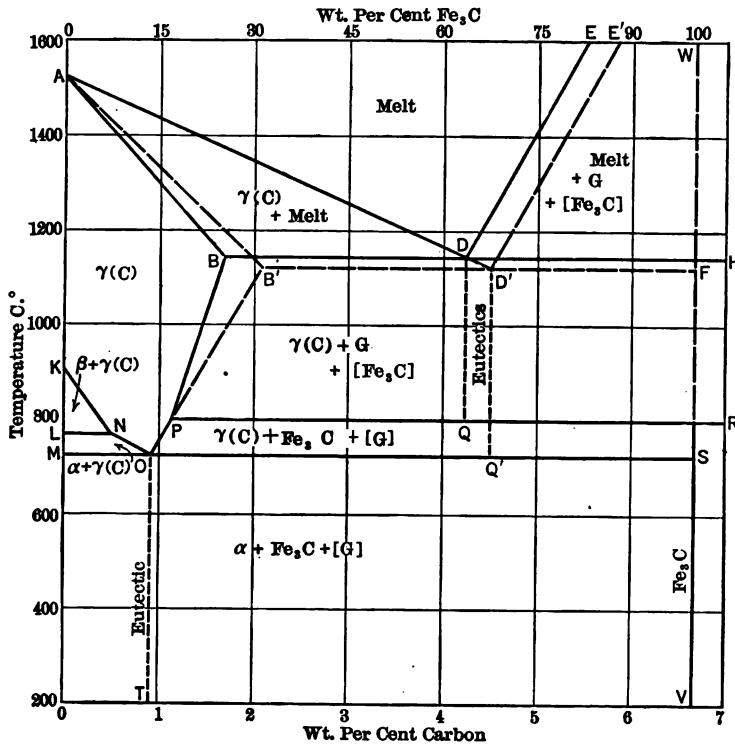


FIG. 79.

it is often convenient to consider the alloy as having the components iron and  $\text{Fe}_3\text{C}$  rather than iron and carbon. Lines with long dashes and symbols in [ ] belong to phases theoretically unstable but often practically permanent or of such long life that they must be considered.

**233.** The phases represented on the diagram, with their names and natures, are as follows. (It should be remembered that there are three allotropic forms of iron. In default of better names these are commonly referred to as  $\alpha$ ,  $\beta$  and  $\gamma$  iron.)

**"Melt"** is the liquid solution of carbon in iron.

$\gamma$  (C) is the symbol for the solid solution of carbon (or some carbide of iron?) in the gamma allotropic form of iron. This solid solution is called **"austenite."**

$\beta$  is the symbol for the beta allotropic form of iron.

$\alpha$  is the symbol for the alpha allotropic form of iron. This is called **"ferrite."**

$\text{Fe}_3\text{C}$  is the carbide of iron, called **"cementite."**

G is crystalline carbon in the graphitic form.

$\gamma$  (C) or austenite is a non-magnetic metal rather similar in engineering properties to copper, but much superior to copper in strength and ductility. The ductility, strength, hardness and shock resistance of austenite all increase with amount of carbon in solid solution.

$\beta$  iron is generally similar to  $\alpha$  but is non-magnetic and is stronger.  $\beta$  iron does not hold carbon in a stable solid solution.

$\alpha$  iron or ferrite is the metal commonly known as iron. It is strongly magnetic, very ductile, tough, of rather low yield-point strength and maximum strength. The change from  $\gamma$  or  $\beta$  to  $\alpha$  iron is sharply marked by the appearance of magnetism.

$\text{Fe}_3\text{C}$ , cementite, is a brittle substance, harder than glass. Its free (primary) crystals tend to occur in the form of thin plates. This crystal form combined with the brittleness of the cementite tends to destroy the ductility of the entire metal containing free cementite crystals.

Graphite occurs in thin plate crystals, black in color and in physical properties of strength, etc., similar to chalk. In gray cast iron the graphite crystals are large enough to be visible to the naked eye.

**234.** At the points  $D$  and  $D'$ , Fig. 79, there occur in the iron-carbon alloys eutectics similar to that already discussed in the lead-tin alloys.  $D$  is the stable eutectic of austenite and graphite.  $D'$  is the unstable eutectic of austenite and  $\text{Fe}_3\text{C}$ . Either or both of these eutectics may be obtained at the end of the freezing of the alloy. With slow cooling the stable eutectic is dominant; with rapid cooling, the unstable eutectic predominates. At  $O$ , Fig. 79, there is a eutectic of a somewhat different class. Eutectics previously discussed have come from the decomposition of a liquid solution. The eutectic at  $O$  is the decomposition of the solid solution  $\gamma$  (C), or austenite, into a

mixture of minute crystals of  $\alpha$  and  $\text{Fe}_3\text{C}$ , or ferrite and cementite. This "solid eutectic" structure of  $\alpha$  and  $\text{Fe}_3\text{C}$  goes under the special name of "**pearlite**." The engineering properties of pearlite are related to those of ferrite and cementite much as the lead-tin eutectic is to lead and tin in Fig. 78. Steels with 0.9 per cent carbon consist normally, at room temperature, entirely of pearlite. With less carbon than 0.9 per cent the normal structure is a mixture of ferrite and pearlite; with more carbon than 0.9 per cent the normal structure is a mixture of pearlite and cementite.

**235.** Iron-carbon alloys with less than 2 per cent of carbon are called steels or wrought irons. The difference between steel and wrought iron is in the manufacturing process. Wrought irons are allowed to begin to solidify in the furnace, where the pasty metal is raked up into balls. In this collection of the pasty metal slag is unavoidably trapped in the metal. Steel is tapped out in the liquid condition from the bottom of the furnace, leaving in the furnace the slag, which floats on the liquid steel. The difference between steels and wrought irons is the absence of slag from the steels. Iron-carbon alloys with more than 2 per cent carbon are cast irons. Commercial cast irons contain from 2.5 to 4 per cent carbon.

**236.** We may now follow the progress of the cooling of an iron-carbon alloy, choosing for a specific case a carbon percentage of 0.3 ("machine" steel). This alloy will begin to freeze at a temperature just under  $1500^\circ\text{C}$ . The crystals formed are of  $\gamma(\text{C})$ . The fact that carbon goes into the solid solution prevents to some extent the actions that have been studied in the freezing of the lead-tin alloys. The last of the melt is used up in the case of the 0.3 per cent carbon steel when the cooling goes below about  $1430^\circ\text{C}$ . Below this temperature, at 0.3 per cent carbon on the line  $AB$ , Fig. 79, this steel is completely solid. From  $1430^\circ\text{C}$ . down to about  $800^\circ\text{C}$ . no further changes occur. The steel is in the austenite condition. At about  $800^\circ\text{C}$ . the 0.3 per cent carbon steel crosses the line  $KN$  of the equilibrium diagram. The  $\gamma(\text{C})$  becomes unstable and spontaneously breaks up into pure  $\beta$  iron and  $\gamma(\text{C})$ , all of the carbon concentrating in this remaining  $\gamma(\text{C})$ . At  $770^\circ\text{C}$ . the line  $LN$  is crossed.  $\beta$  iron disappears and  $\alpha$  appears. Some  $\gamma(\text{C})$  remains, now quite rich in carbon. With the appearance of  $\alpha$  the steel becomes magnetic. As

the cooling continues the remaining  $\gamma$  (C) decreases in amount and becomes richer in carbon till its composition reaches the point O. At that point  $\gamma$  (C) forms both  $\alpha$  and  $\text{Fe}_3\text{C}$  crystals simultaneously, making pearlite. The point O is at a temperature a little above  $700^\circ\text{C}$ . Below  $700^\circ\text{C}$ . the 0.3 per cent carbon steel has no further changes. It comes down to ordinary temperatures as a mixture about two-thirds ferrite and one-third pearlite.

**237.** It may be seen from this that the changes which occur in the solid steel at temperatures around  $700^\circ$  to  $800^\circ\text{C}$ . are even more important in the determination of the final structure and properties of the metal than changes during the freezing of the metal. Hence the temperatures at which changes occur in the solid steels and cast irons have come to be called **critical temperatures**. Specifically the critical temperature is that of the lines *LNOS*, Fig. 79, above which  $\alpha$  is absent and below which  $\alpha$  is normally present.

## CHAPTER XIII

### THE SHAPING OF STEEL AND THE CONTROL OF FINAL PROPERTIES DURING THE SHAPING PROCESSES

**238.** The shaping of steel into the form desired for use may be done either by casting directly into the final form or by casting a rough block, called an **ingot**, and shaping the ingot by mechanical processes. The latter is, by far, the more common method.

The steel ingot is made by pouring liquid steel into a cast-iron mold. This mold is usually rectangular in its horizontal cross-section, open at top and bottom, and slightly tapered from the bottom toward the top so that it may be "stripped" or pulled off from the ingot. The part of the liquid steel in contact with the heavy, comparatively cold walls of the mold and the pouring floor, freezes almost immediately. As soon as the outer parts of the ingot are strong enough to hold the still liquid interior, the mold is pulled off from the ingot. Small ingots are allowed to cool down completely and are reheated for forging. Large ingots are put into a "soaking pit" (a heated chamber) to cool slowly until they are solid throughout and are then taken, with no further cooling, to the rolls or forges.

**239.** Certain defects of the finished steel have their origin during the freezing of the ingot. The causes are **pipes**, **blowholes** and **segregation**. Steel contracts in volume considerably in changing from liquid to solid. The outer parts of the ingots are rapidly cooled and formed by contact with the ingot mold. Across the top of the liquid steel in the ingot mold, a crust forms by loss of heat to the air. These solid outer parts surround, at the time, a liquid interior. As this liquid interior becomes solid there is not enough material to fill the entire volume within the crust. The result is a cavity in the top and center of the ingot as shown in Fig. 80. This cavity is called a pipe. A pipe is usually lined with impurities so that it cannot be successfully welded up during the rolling or forging of the ingot. There would result a defect (practically a crack) in the finished piece.

We must either prevent the formation of the pipe or discard the upper end of the ingot. The latter is the usual practice. It is called "**cropping**." The amount of the cropping varies from 10 to 30 per cent of the ingot. The amount of cropping is often

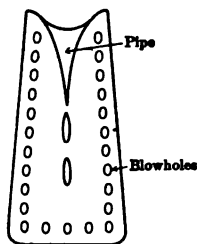


FIG. 80.

specified in contracts between steel maker and purchaser. Piping can be prevented by keeping the upper end of the ingot fluid until the lower part has solidified. This may be done by direct heating by flames or by the use of thermit. Both of these devices have been worked out into commercial processes reducing the amount of cropping to less than 10 per cent.

**240. Blowholes** are due to gases which have been dissolved in the liquid steel. The solid steel will hold in solution only a fraction as much of these as the liquid steel will hold. Hence just after the metal begins to freeze gases are evolved. They form a layer of bubbles more or less parallel to the surfaces of the ingot, as shown in Figs. 80 and 81. Small blowholes can be welded up in hot working of the steel. Large blowholes do not weld up perfectly, but remain as cracks parallel and near to the finished surfaces of the metal. Blowholes may be prevented or minimized. By proper treatment of the liquid steel in the furnace the amount of gases absorbed may be kept down. After tapping the steel from the furnace chemicals may be added to the liquid steel to combine with the gases. Such chemicals are silicon and aluminum. Pouring must always be done at a correct temperature. Pouring too hot or too cold gives larger blowholes than an intermediate temperature.

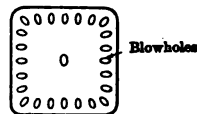


FIG. 81.

**241. Segregation** means that the chemical composition of the ingot as to percentages of carbon, phosphorus, sulphur, etc., is not uniform throughout the ingot. The outer parts of the ingot freeze so quickly in contact with the mold that they have the same analysis as the liquid steel. The core of the ingot which freezes last splits more or less into two layers, the lower one relatively poor in carbon, phosphorus, and sulphur and the upper one relatively rich. This is illustrated in Fig. 82 by the shading of the figure. The lean lower central portion of

of the average amount of which upper central portions contain in extreme cases over twice as much carbon as the lower portions. The tendency of segregation is toward a lower density with increasing carbon content. Because of the high viscosity of the liquid metal, the segregation is either caused by the difference in density  $\gamma$  (C,P), or by the difference in density of any temperature range. This is especially true in the case of carbon or phosphorus segregation. This is the case of a line of segregation in the upper portion of an ingot, as shown in Fig. 79.  $AB$  is the line of segregation when the austenite is in equilibrium with the melt. The simultaneous composition of the austenite is similar to that of carbon. As the ingot is freezing, the formation of the upper portion of the ingot is freezing, helps in the segregation of the upper portion of the ingot. The upper portion of a finished piece is much too hard and brittle. The steel is of low average carbon content about 0.10 to 0.15 per cent. This is not a commercially serious matter, because the ingot will still have sufficient carbon content (0.5 to 1.5 per cent) to be used. It must be controlled.

Segregation. If the freezing is too rapid, there is no time for segregation. On the other hand, if the cooling is too slow, the carbon would diffuse until all the carbon is in the upper portion. The very slow cooling of the ingot by a natural correction of segregation has been found to lessen segregation.



FIG. 82.

gation in large ingots. It is a fortunate fact that with the exception of railroad rails we do not need to use large ingots with high-carbon steels. ("High carbon" means above 0.5 per cent.)

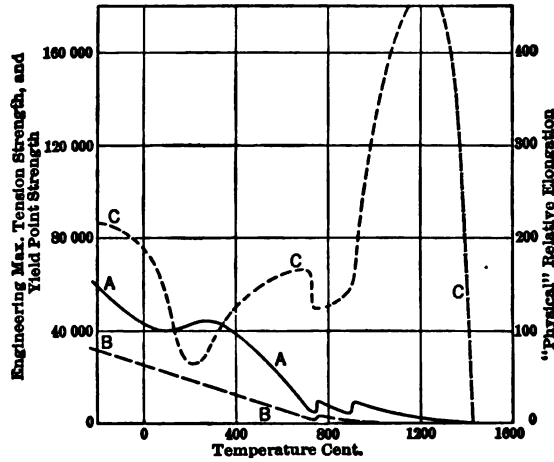


FIG. 83.

242. From the soaking pit or the reheating furnace the ingot goes to a set of "roughing" rolls. In a few passes back and forth through

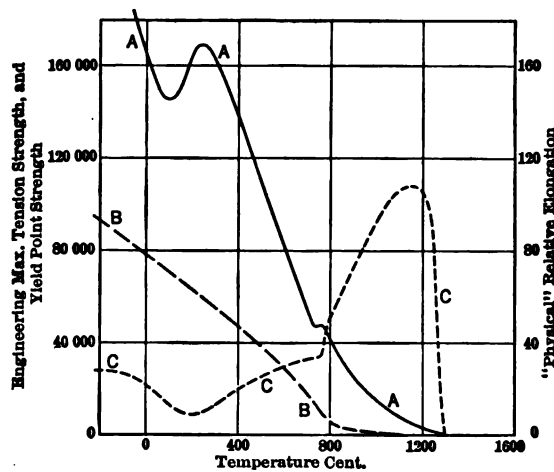


FIG. 84.

these rolls the ingot is greatly reduced in cross-section and increased in length. Its cross-section shape remains rectangular. Shaping



of the piece now begins. If the shaping is done by rolling as for rails, I-beams, channels, etc., the rolls in each pass slightly reduce the cross-section and increase the length of the piece and at the same time bring it nearer to its finished form and size. For more complicated shapes the piece may be formed by forging under hammers or presses. The heavy work of shaping the piece is finished while the piece is still above a red heat. With large pieces of material the heat used may be that from the original ingot. Small work cools more rapidly and requires reheating.

**243.** The reason why the heavy working of the steel is done at high temperatures is shown in Figs. 83 and 84, which represent approximately the variation of the properties of the metal as a function of temperature. In each figure the curve *A* is the engineering maximum tension strength of the metal, *B* is the yield-point strength, and *C* is the "physical" relative elongation. The "physical" relative elongation is used instead of "ductility" because it more nearly corresponds to the increase in length of a piece being rolled out. The curves *C* show what extension a metal would stand if pulled out to break in one loading. Above red heat the metal recovers between passes through the rolls so that the actual extension possible by repeated loadings is very great. Fig. 83 is for steel with no carbon and Fig. 84 for steel with 1.00 per cent carbon. For other carbon contents similar curves may be interpolated from these two sets in proportion to carbon content. Inspection of the curves shows that working becomes easier as the temperature rises, because the strength of the material decreases; and that large changes in the shape of the metal, without breaking the piece, are possible only at temperatures above 800° C. On the other hand, large changes of shape of the metal cannot safely be made at temperatures higher than that of the peak of the elongation curves *C*, near 1200° C.

**244.** The upper and lower limits of temperatures for heavy working of steels vary with the carbon content of the steel. Other chemical elements than carbon have little importance in this connection unless present in very large amounts. Ductility of the steel is entirely lost at and above the line *AB*, Fig. 79. Ductility is low until the temperature has fallen appreciably below *AB*. Because of this, and because the temperature of the piece is not quite uniform from

one part to another, the upper limit of working temperatures for steels is a line parallel to  $AB$  and somewhat more than  $100^{\circ}\text{C}$ . below  $AB$ . The lower limit of temperatures for heavy working is fixed by the sharp drop of ductility and rise of strength near  $800^{\circ}\text{C}$ ., Figs. 83 and 84. The temperatures at which these changes occur vary with carbon along the line  $KNOP$ , Fig. 79. Heavy working must stop at a temperature higher than that of  $KNOP$  by  $50^{\circ}$  to  $100^{\circ}\text{C}$ . It will be noticed that these limits of temperature for heavy working are simply the boundaries of the austenite or  $\gamma$  (C) field, Fig. 79. Moderately heavy working may also be done between temperatures of  $900^{\circ}$  and  $1100^{\circ}\text{C}$ . with carbon content as high as 2 per cent, though this is beyond the right-hand edge of the austenite field. Elements such as tungsten, chromium, nickel, etc., generally raise both the lower and upper limits of the working temperatures.

**245.** The methods of working differ somewhat in their effects on the material. Rolling and pressing cause deformations which go throughout the body of the material, working the inner parts nearly as well as the outer parts. Pressing is somewhat better than rolling in this respect. Hammer blows have rather local action. Unless the hammer is very heavy compared to the piece of material being forged, the working of the material is incomplete in the inner parts of the piece.

**246.** After the end of the shaping of the steel in hot working, the still red hot pieces are run out onto the "cooling beds" to cool down to ordinary temperatures. It is essential that during this cooling all parts of a piece should cool uniformly and simultaneously. Uneven cooling causes uneven shrinkage, which sets up heavy "cooling stresses" in the metal. Uniform cooling is usually secured by making the cooling very slow, putting considerable numbers of hot pieces near each other on the cooling beds.

**247.** During the heavy working of the hot steel, pipes or blow-holes may cause the piece to split in the rolls or under the hammer. Another cause of trouble in hot working of steels is the presence of too much sulphur (and oxygen) in the steel. The  $\text{FeS}$  (and  $\text{FeO}$ ) melts at a temperature slightly above  $900^{\circ}\text{C}$ . so that little globules of liquid  $\text{FeS}$  are mixed with the solid steel at the very temperatures where we desire to do the heavy working. This makes the steel low

in ductility or relatively brittle so that heavy working is not possible with high sulphur. The steel is said to be "**hot short.**" The addition of manganese in making the steel results in the formation of MnS instead of FeS. The melting point of MnS is above  $1100^{\circ}\text{C}$ . By keeping the amount of sulphur under 0.10 per cent and using enough manganese to keep the sulphur in the form of MnS, hot shortness is avoided.

**248.** While light working of the steel may continue between  $900^{\circ}$  and  $550^{\circ}\text{C}$ ., where the red color of the hot metal disappears, it is decidedly dangerous to do even light working at temperatures between  $550^{\circ}$  and  $150^{\circ}\text{C}$ . The reason for this may be seen in Figs. 83 and 84 in the very decided minimum in the elongation curves at about  $200^{\circ}\text{C}$ . This is the region of the so-called "**blue heat.**" The brittleness of iron and steel at this temperature is called a "**blue heat brittleness.**" At ordinary temperatures around  $20^{\circ}\text{C}$ . the ductility is again rising and wrought irons and steels of low carbon content will stand appreciable amounts of light working.

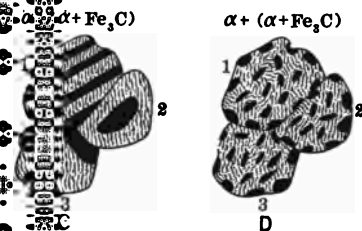
**249.** At ordinary temperatures steel is made brittle against shock, but not against slowly applied loads, by the presence of too much phosphorus. The amount of phosphorus which can be tolerated depends very much on the amount of carbon present. Shotgun barrels with very low carbon content contain 0.4 per cent P and give no trouble from brittleness. (Shotgun barrels are made by repeated welding and high phosphorus makes this welding easy.) Railroad rails and tool steels must have the phosphorus considerably under 0.10 per cent in order to avoid "**cold shortness.**"

**250.** The working of the steel has another purpose than that of mere shaping of the material. At the same time that the working is changing the shape of the material it also changes the internal structure. The average crystal size is made much smaller or the "**grain is refined.**" Fine grain in steel goes with the best of engineering properties. Coarse grain means poor quality. **The refining of the grain during working is quite as important an object of the working as is the shaping of the piece.**

A tool steel of 0.9 per cent carbon in the unhardened condition shows the following properties at different finenesses of grain. The finest grain possible in the soft condition of the steel is that called

# RUCTION

structure gives a strength of 70,000 pounds per square inch with a ductility of 10 per cent. This structure, which is easily obtained, gives in this condition a strength of 70,000 pounds per square inch with a ductility of 10 per cent. If the steel is heated for several hours at 1,000° F., the structure becomes gradually coarser, passing from "fine pearlite" to "segregated mass pearlite." This coarse pearlite gives a strength of 60,000 pounds per square inch with a ductility of 5 per cent. The following table illustrates the range of properties which can be obtained merely of the size of its structure. The range of structure possible is apparent. The structure of the steel is in the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  while the steel is in the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  of the steel ceases at temperature 1,000° F. The steel then cools naturally and the structure becomes  $\alpha + (\alpha + \text{Fe}_3\text{C})$ . Practically all steels pass as they cool through the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  and the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  is undergone by the crystals. Fig. 85A shows a group of



these are in critical temperatures. The structure of the steel is in the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  while the steel is in the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  of the steel ceases at temperature 1,000° F. The steel then cools naturally and the structure becomes  $\alpha + (\alpha + \text{Fe}_3\text{C})$ . Practically all steels pass as they cool through the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  and the condition of  $\alpha + (\alpha + \text{Fe}_3\text{C})$  is undergone by the crystals. Fig. 85A shows a group of

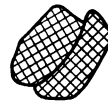
forms as in Fig. 85B, 2 and 3. If the austenite crystal is homogeneous it is least stable at its boundaries, and ferrite forms as in 3. The austenite crystals often are not homogeneous internally. During the freezing of the metal the successive layers of molecules building up the austenite crystals are progressively richer in carbon. (See line  $AB$ , Fig. 79.) If sufficient chance has not been given after freezing for the carbon to diffuse to uniformity, the core of each austenite crystal will be lower in carbon content than the outside. The line  $KNO$ , Fig. 79, shows that the lower the carbon content of the  $\gamma$  (C), the higher is the temperature at which it begins to form  $\beta$  or  $\alpha$ . A "cored" crystal of austenite may therefore form ferrite at its center as in Fig. 85B, 2, instead of at the outside as in 3. As the cooling progresses through the field of  $\alpha + \gamma$  (C) more and more ferrite is deposited upon the original ferrite nuclei, while the remaining austenite decreases in amount and becomes richer in carbon until it reaches the temperature and composition of the point  $O$ , Fig. 79. Further cooling causes the last of the austenite to decompose to an intimate mixture of minute cementite and ferrite crystals called pearlite. After this change has occurred the structure of Fig. 85B is changed to the condition of Fig. 85C. No further changes will occur in cooling to ordinary temperatures.

The "grain" of the cold steel is one of the groups of ferrite and cementite crystals such as 1, 2 or 3, Fig. 85C. The form and size of this "grain" are simply the form and size of the single austenite crystal from which the whole was formed by going through the stages in Fig. 85 A, B and C. This grain is readily picked out in the microscopic examination of the steel.

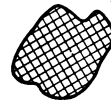
**252. The rate of cooling through the critical temperatures determines the size of the ferrite and pearlite patches in the grain of the steel.** If the cooling is comparatively slow only a few ferrite nuclei form in the austenite crystals and the ferrite builds up on these nuclei to form few and large ferrite crystals. This is shown in Fig. 85B and C. If the cooling is accelerated (but not made sharp enough to harden the steel) many more ferrite nuclei are formed and none of them grow to large size. Even the ferrite and cementite crystals of the eutectic structure are finer than with slower cooling, yielding sorbite instead of pearlite. Fig. 85D shows the result of more rapid

## RUCTION

ation of sorbitic to pearlitic visible grain; the outlines completely lost. formation from liquid metal solid metal, but also with temperature. The rate of with the temperature. In ceptible at temperatures as h is not perceptible below mportant in steels until the ter a few hours of exposure steel goes through a change teel goes through a change ed pearlite structure. An ure will bring the structure ee section 250.) When the atures or into the austenite and possibly serious. The growth es instead of hours or days. grow in the solid metal finds e around the molecule which crystals. Within the crystal, etric pattern, which may be d at each of the intersections ange of action of the directed



C



D

compared to the size of the considered to be surrounded, as by a directed field of force, into a parallel crystal struc-

ture. When the adjacent material is in the liquid condition, this action can easily be shown. Fig. 86B is a photograph of a primary crystal of a solid solution of antimony and tin surrounded by a eutectic structure. Part of the crystals of the eutectic are of the same material as the big primary crystal. The alignment of the crystals of the eutectic by the action of the directed field of force from the primary crystal is very evident. The eutectic changed from liquid to solid in the presence of the primary crystal. The orientation of the crystals of the eutectic was determined by the directed field of force of the big primary crystal.

Suppose now two crystals of the same kind in the solid metal are originally related, in their crystal patterns, as shown in Fig. 86C. The tendency of these two patterns is to become parallel to each other. Every molecule is vibrating about its mean position. This vibration constitutes the heat motion of the molecules in the solid. With rise of temperature the amplitude and energy of this vibration increase. This increase of the heat motions of the molecules reduces the rigidity of direction of the crystal patterns, and at a sufficiently high temperature the two patterns will suddenly coalesce. The boundary between the crystals disappears. The result is a single crystal as in Fig. 86D.

255. The smaller the crystals the lower is the temperature at which they will combine. With increase in size the ability to combine to still larger sizes decreases. Crystal growth depends on time and temperature somewhat as shown in Fig. 87. At low temperatures growth is negligible for any length of time however large. At high temperatures the initial growth is rapid. Each added length of time has less and less effect, in some inverse proportion to the total preceding time.

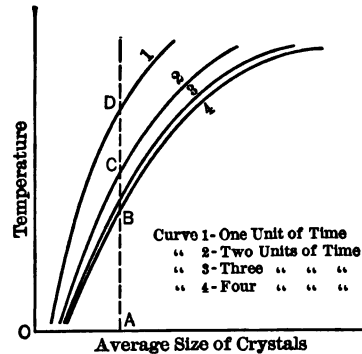


FIG. 87.

Let it be supposed that the largest crystal size found by practice to be permissible in material of good engineering properties is shown by OA, Fig. 87. This size will be reached in a fairly long exposure

of the metal at the temperature *B*, in half the same time at temperature *C*, and in a quarter of the time at temperature *D*. This illustrates the necessity of increasing care about time of heating as the temperatures are higher.

256. When a certain crystal size has been attained by growth, the crystals do not "ungrow" in cooling, but retain the size they had reached. It has been shown in Fig. 85 and its discussion, section 251, that the grain size of the finished steel depends on the size of the austenite crystals just before they reach the critical temperatures in cooling. (This is somewhat subject to changes by rate of cooling.) A steel which has been made coarse grained by growth of crystals is called "overheated." Overheating is likely to occur at any temperature within the austenite field.

257. In a steel ingot the size of the crystals of austenite formed during the freezing of the metal is about one-half inch in diameter.

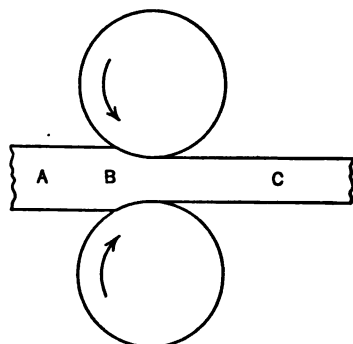


FIG. 88.

The ingot is repeatedly passed through rolls as shown in Fig. 88. At each pass the cross-section area is considerably reduced. There are fifteen to thirty passes through the rolls before the pieces emerge in finished form and size. At each pass the austenite crystals are mechanically broken into smaller crystals. Fig. 89 illustrates how this mechanical breakdown of the crystal occurs. A, B and C of Fig. 89 correspond to A, B and C of Fig. 88. A, Fig. 89, may

represent the size of an austenite crystal approaching the rolls. B shows the formation of shear planes through the crystal as the metal enters the rolls. In passing through the rolls the portions of the crystal are slipped across each other on these shear planes to so large an extent that the crystal is effectively broken into a group of smaller crystals. This group is shown in C. The amount of mechanical breakdown of crystals, during the working of the steel, can be measured by comparing the initial and final crystal sizes. The grain size of the finished piece considered good (that is, not coarse grained)



does not exceed 0.005 inch. As this is only  $\frac{1}{1000}$  of the initial diameter of austenite crystals in the ingot, the volume change by mechanical breakdown is in the ratio of  $1000^3$ , or  $10^9$ , to 1. This may at first seem an impossibly large number. Under the minimum assumption that there are fifteen passes through the rolls and at each pass each crystal is made into two, the reduction ratio would be  $2^{15}$  to 1, or  $10^{4.5}$  to 1. It will be seen from this that the reduction given above is entirely possible. This outlines the method by which, during the working of the steel, we control not only the form and size of the fin-

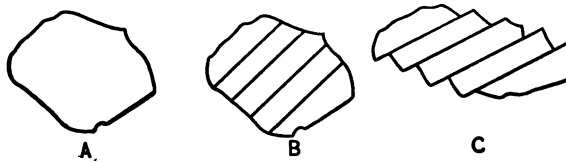


FIG. 89.

ished piece, but also the grain size of the finished metal and consequently the engineering properties of the metal.

258. Because we must control the grain size during the shaping of the piece, we must have a definite relation of cross-section area of the original ingot to the cross-section area of the finished piece. The greater this ratio the greater is the number of passes required through the rolls (or the amount of working) and hence the better the grain size of the finished piece. If the area of the ingot cross-section is too small the working will be insufficient to secure fine grain. By experience we compromise on a definite size of ingot as sufficiently large to give enough working for fine grain and not so large as to make an excessive amount of working. Reducing the grain size below 0.005 inch does not greatly improve the engineering properties of the metal.

259. In Fig. 89C, it will be seen that the parts of a crystal just broken down by heavy working will have their internal crystal patterns very closely parallel and will, therefore, readily grow together again. This growth of crystals after working is peculiarly rapid and must always be guarded against. To be effective in reduction of grain size, the working of the metal must be continuous. Also, since the growth of crystals in steel is comparatively rapid at any temperatures above the critical, the heavy working for reducing the

**grain size must continue until the temperature of the piece has fallen close to the critical temperatures.** Figs. 83 and 84 show, however, that the working must be progressively lightened as the temperature falls. If all of the heavy working of a piece of steel were completed at 1000° C. and the piece then cooled in air, the piece would be coarse grained from the growth of the crystals after the end of the working, because the air cooling allows sufficient time for the growth. A more rapid cooling, as by a spray of water, bringing the temperature below the critical in a few seconds, would prevent crystal growth after the end of the working. The finishing temperature of the working must be close to the critical temperature, or accelerated cooling must be used. The accelerated cooling should extend only down to the critical temperature; below the critical temperature the cooling should be normal.

**260.** Occasionally a piece is brought to its finished size and shape without having secured fine grain within the metal. Further working for refining the grain is out of the question. The resource is a heat-treatment method for refining the grain. The outline of this heat treatment is as follows: The start is with a coarse-grained steel with a structure of ferrite and pearlite. The steel is heated as rapidly as possible, consistent with uniform heating of the piece, to a temperature from 50 to 100 degrees above the critical temperature. In passage through the critical temperatures, pearlite and ferrite become unstable and spontaneously decompose with formation of austenite,  $\gamma$  (C). At the instant of formation the austenite crystals are small, but they immediately begin to grow. Hence the piece should be kept above the critical temperature only long enough to be sure that all the pearlite and ferrite have decomposed. The time for this depends on the size of the piece, on the proportions of ferrite and pearlite, and on the size of the original coarse grain. The pearlite changes almost instantaneously to austenite at the critical temperature. The ferrite is slowly absorbed by the austenite crystals formed from the pearlite. As soon as ferrite has disappeared the piece should be cooled. In cooling, the austenite decomposes to ferrite and pearlite. Again the new crystallization, at the instant of its formation, is finer than the preceding structure. The cooling, through the critical temperatures, should be rapid enough to secure a structure of the type shown in

Fig. 85D, rather than C. Air cooling of small pieces will do this. Accelerated cooling to about  $650^{\circ}\text{C}$ . (a dull red) followed by air cooling to room temperature would be the ideal. It may be advisable to go through the heat treatment twice. Heat-treatment refining of the grain is quite as effective as the mechanical method.

**261.** The preceding discussion, in Sections 250 to 260, of crystal growth and the mechanical and heat treatment methods of control of grain size, is summarized graphically in Figs. 90 to 93. These figures all have temperature as ordinate and grain size as abscissa. The critical temperature runs across the middle of each diagram. Arrowheads mark the direction of the temperature change.

Fig. 90 shows the mechanical reduction of grain size as it occurs in ordinary hot working, such as rolling or forging. Starting with coarse-grained material (generally ingot or billet) at *A*, there is heavy working and rapid reduction of grain size from *A* to *K*. Along *KL* the working ceases momentarily, and there is a rapid partial recovery or growth of the austenite crystals back toward their initial size. *KL* might represent the period of passage of the piece from one "pass" to another in the rolls of a rolling mill. *LM* is a second period of working; *MN* a second rest; and so on to *B*. The working must be heavy enough, and nearly enough continuous, to keep ahead of crystal growth. Working stops at *B*, somewhat above the critical temperature. From *B* to *C* the piece cools, with some crystal growth; but growth of austenite crystals is not very rapid just above the critical temperature. In slow cooling, decomposition of austenite into ferrite and pearlite begins at *C*, or the critical temperature, and continues from *C* to *D*. The ferrite and pearlite structure has practically no crystal growth in air cooling of the piece from *D* to *E*, at room temperature.

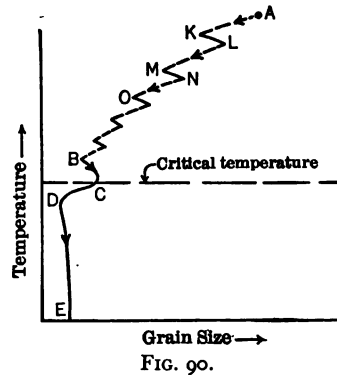


Fig. 91 shows in contrast to each other the effects of different rates of cooling from the lower edge of the austenite field, just above the critical temperature. *BCDE* represents the air cooling of a small

piece of steel. Because of the moderately rapid rate of temperature fall the point *C*, the beginning of austenite decomposition, comes a bit below the critical temperature. Along *CD* the austenite decomposes to the very fine mixture of ferrite and cementite called sorbite.

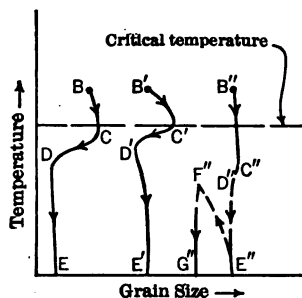


FIG. 91.

*B'C'D'E'* represents the air cooling of a large piece of steel, or the furnace cooling of a small or large piece. There is growth of austenite from *B'* to *C'*, formation of pearlite from *C'* to *D'*, and some growth of the ferrite-pearlite structure below *D'*. The final structure, at *E'*, is not refined as compared with the austenite crystal size at *B'*; in the more rapid cooling shown in *BCDE* a refining does occur. A very

rapid cooling, quenching of the steel in water from the austenite condition at *B''*, gives the austenite persisting to *C''*, then changing along *C''D''E''* to the hard martensite, with little refining. Hardening of steel involves comparatively little refining of grain. If the hardened steel, in the martensitic condition, is reheated toward red heat and cooled again (tempered) refining does occur, indicated by *E''F''G''*. The final result in the hardened and tempered steel, at *G''*, bears practically the same relation to *B''* that *E* does to *B*, in the air cooling of a small piece.

Fig. 92 shows what happens essentially in the **overheating** of steel. The steel is heated, probably slowly, to so high a temperature and for so long a time that considerable crystal growth of austenite occurs along *DE*. Then slow cooling along *EFGH*, with no working (or insufficient working), leaves the structure finally, at *H*, considerably coarser in grain than the original. Steel should never be heated above the point *D* (about 50 to 100° C. above the critical temperature) unless such heating is to be followed by refining of the grain by working or by heat treatment.

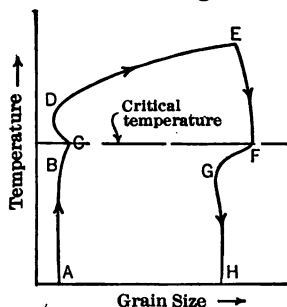


FIG. 92.

Fig. 93 indicates the process of heat treatment refining of the grain.

Starting at *A* with coarse-grained steel, the piece is heated as rapidly as possible without unevenness of temperature throughout its parts, and consequent expansion stresses, along *ABCD*. From *C* to *D* the structure of ferrite and pearlite changes to austenite, with refining during the decomposition of one set of crystals and formation of another. Heating beyond *D* would start growth of austenite. This is avoided by taking the piece out of the furnace as soon as the austenite formation is complete (magnetism has disappeared) and cooling fairly rapidly, though not fast enough to harden the steel. In this cooling shrinkage stresses, due to uneven temperatures of different parts of the steel, must be avoided. From *E* to *F* ferrite and sorbite form, with refining, as in Fig. 91, *BCDE*. The cold steel, at *G*, Fig. 93, is considerably finer in structure than the original at *A*. The process may be repeated along *GC'D'E'F'G'*, with further refining; or since the refining is due to crystal decomposition and formation around the critical temperatures, the essential part of the action is in *BCDEFC'D'E'F'*, etc. Usually one passage up and down through the critical region is sufficient for commercial purposes. Heat treatment for refining of the grain is necessary for steel castings if they are to have properties to make them compete with forgings; it is to be recommended for many forgings also if there is need for the utmost of strength and toughness of which the steel is capable.

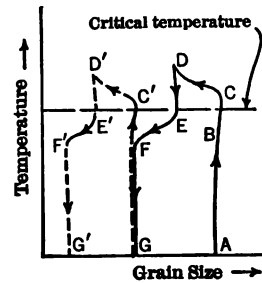


FIG. 93.

**262.** Of the methods of shaping metals, casting is generally cheaper than working. The casting of steel is peculiarly difficult. Dry sand molds must usually be used instead of "green" sand, because of the evolution of steam which would occur with moist sand. (The pouring temperature for steel casting is considerably higher than for cast iron, and the steam evolution is more vigorous.) Cores must be very easily crushed by the metal as it shrinks, or the casting will be broken. Shrinkage is large both during freezing and afterward. Shrinkage during freezing requires the use of large sprues, risers and feeding heads to supply liquid metal to the casting as it freezes. After solidification the casting shrinks rapidly in cooling until the critical

temperature is reached. The change from austenite to ferrite and pearlite involves a considerable increase in volume, or expansion. Below the critical temperatures there is uninterrupted shrinkage to ordinary temperatures. The net shrinkage from the pattern to the cold casting is about  $\frac{3}{8}$  of an inch to the foot. The large shrinkage makes it necessary to have uniformity of thickness in different parts of the casting. If one part cools faster than another its shrinkage would be different and severe internal stresses would be set up. Steel castings require much more careful design of patterns than do cast-iron castings.

During the freezing the austenite crystals formed are large. The internal stresses in cooling, especially those due to irregular expansion and shrinkage near the critical temperatures, cause the austenite crystals to change to ferrite and pearlite in such patterns as in Fig. 85, B and C. The structure of Fig. 85C, 1, often occurs. The ferrite in the final structure is in great plates. A steel casting therefore, when it comes from the sand, is weakened by internal stresses, and is of poor quality of metal, coarse grained, and with a very bad distribution of the ferrite.

**263.** To remedy these defects, there is used a heat treatment called **annealing**. The purpose of annealing is to remove the shrinkage stresses and to refine the grain. The treatment is essentially that outlined above for refining the grain of steel. Because castings are usually complicated in form, the only practicable way of keeping uniformity of temperature throughout the casting in heating or cooling is to proceed slowly. Uniformity of temperature throughout the casting must be had even at the cost of less effective heat treatment. Good castings, properly annealed, are now made quite comparable in properties with forgings.

**264.** It is not possible in hot working of a metal to secure exact sizing of the finished piece, because of the rather uncertain amount of shrinkage in cooling after the working. For exact shaping and sizing either machining or **cold rolling** must be used. Cold rolling cannot produce large changes in shape and size of the piece, but it does enable finishing to exact sizes and smooth surfaces. At the same time the metal in the surface of a cold-rolled piece is hardened and has its yield point raised by the cold working. This makes cold rolling an ideal method for finishing shafting.

Cold rolling breaks down the ferrite crystals of a steel and so has the effect of refining of grain. Reheating to dull red,  $650^{\circ}\text{C.}$ , will completely recrystallize the ferrite, putting the metal back in the condition it had before cold working. Reheating to lower temperatures than dull red causes first a softening, a loss of the hardness due to the cold working, and secondly, a partial recrystallization of the ferrite.

## CHAPTER XIV

### ENGINEERING PROPERTIES OF NORMAL CARBON STEELS AS FUNCTIONS OF THE CARBON CONTENT; EFFECTS OF ELEMENTS OTHER THAN CARBON

265. The steels used in ordinary commercial work are "carbon" steels, that is, alloys of iron and carbon, so slightly modified by other chemical elements that the properties may be considered as determined by the iron and carbon. The remaining elements may be regarded as impurities. When other chemical elements than iron and carbon are present in large enough amounts to control the properties, the steel is called an alloy steel.

266. The word **normal** refers to the condition of the steel. The special meaning used here implies that the steel was hot worked by forging or rolling and then cooled on the cooling beds in the customary commercial way. Normal then means the condition in which the steel would be purchased commercially with the properties unmodified by cold working or special heat treatments. The greater proportion of steel used in engineering structures is used in the normal condition.

267. Considering the effects it produces on the engineering properties, the amount of carbon present in steels is astonishingly small. The carbon content of even the hardest tool steels does not exceed 1.5 per cent by weight. The amount of carbon is measured in hundredths of one per cent. A peculiar practice has now become commercially common of speaking of each  $\frac{1}{100}$  of one per cent of carbon as a "point" of carbon. For example, a "twenty point" carbon steel means a carbon content of 0.20 per cent.

268. The reason for the great effect of carbon on the properties of the steel is found in the form in which the carbon is present and in the structures which it causes. In normal steels, carbon is present as the chemical compound  $\text{Fe}_3\text{C}$ . The weight of this compound formed from one unit of carbon is fifteen times the weight of the carbon. A normal steel with one per cent of carbon has really fifteen per cent of its weight as  $\text{Fe}_3\text{C}$ .  $\text{Fe}_3\text{C}$ , or cementite, is a brittle substance, harder than glass, and decidedly strong, especially against shear.



269. The structures of the normal steels are determined as they cool through the critical temperatures. Fig. 94 shows on a larger scale the portion of the equilibrium diagram (Fig. 79) with which we are at present concerned. At the end of the hot working the steel is in the austenite,  $\gamma$  (C), condition. The way in which the final structure of normal steels originates in passage through the critical temperatures will be examined for three different compositions.

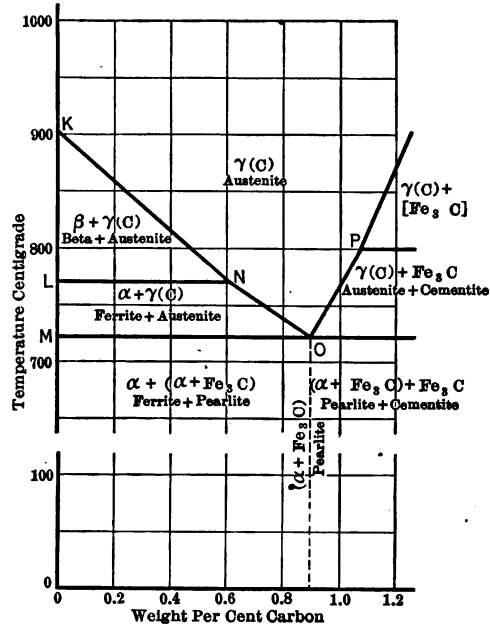


FIG. 94.

A steel of 0.2 per cent carbon remains as  $\gamma$  (C) in its cooling until the temperature reaches about  $870^{\circ}$  C. At that temperature, there appear spontaneously within the  $\gamma$  (C) crystals nuclei of pure  $\beta$  iron. As the temperature falls from  $870^{\circ}$  to  $770^{\circ}$ , the size of the  $\beta$  crystals increases, until at  $770^{\circ}$  in this steel about two-thirds of the whole structure is  $\beta$  crystals. The remainder of the structure is still  $\gamma$  (C). This remaining  $\gamma$  (C) holds in solid solution all of the carbon content of the steel. At  $770^{\circ}$  the carbon content of the remaining  $\gamma$  (C) is about 0.6 per cent. The composition and temperature of the remaining  $\gamma$  (C) has gone down along the line KN, Fig. 94. At  $770^{\circ}$  the  $\beta$  iron spontaneously changes to  $\alpha$  and the steel becomes magnetic. With further cooling the  $\alpha$  crystals grow, the amount of  $\gamma$  (C) decreases and the concentration of carbon in the remaining  $\gamma$  (C) increases. The composition and temperature of the remaining  $\gamma$  (C) move along the line NO. Above the point O the remaining  $\gamma$  (C) of this steel (0.20 per cent carbon) has been decomposing with formation of  $\beta$  or  $\alpha$  iron. At the point O the decomposition of the  $\gamma$  (C) yields

simultaneously  $\alpha$  and  $\text{Fe}_3\text{C}$ . The molecules of these two substances will not build up together into a solid solution. In their attempt to crystallize each interferes with the growth of the crystals of the other. Therefore the  $\gamma$  (C) which decomposes at the point  $O$  produces a peculiar intimate mixture of very minute  $\alpha$  and  $\text{Fe}_3\text{C}$  crystals. This eutectic mixture is called pearlite. [The symbol for pearlite is  $(\alpha + \text{Fe}_3\text{C})$ .] The formation of the pearlite is the last change in the alloy structure as the metal cools. The normal 0.20 per cent carbon steel will have a structure containing about 20 per cent of pearlite intermixed with about 80 per cent of large primary ferrite,  $\alpha$ , crystals.

A steel of 0.8 per cent carbon content will remain as  $\gamma$  (C) down to a temperature of about  $740^\circ\text{C}$ . At that temperature ferrite nuclei will form in the  $\gamma$  (C) crystals. These nuclei will grow to small  $\alpha$  crystals while the remaining  $\gamma$  (C) changes in temperature and composition along  $NO$ , Fig. 94, to the point  $O$ . At  $O$  the last austenite changes to pearlite. This steel will consist, in the normal condition, almost entirely of pearlite. The small  $\alpha$  crystals will amount to less than 10 per cent of the whole structure.

A steel of 1.00 per cent carbon will remain as  $\gamma$  (C) to a temperature of about  $770^\circ\text{C}$ . The  $\gamma$  (C) decomposition in this case differs from the two preceding cases in that the first decomposition product is  $\text{Fe}_3\text{C}$ . The cooling of the steel from  $770^\circ$  to  $720^\circ$  produces growth of the  $\text{Fe}_3\text{C}$  crystals, while the remaining  $\gamma$  (C) changes in composition and temperature along  $PO$ , Fig. 94. At  $O$  the last  $\gamma$  (C) changes to  $(\alpha + \text{Fe}_3\text{C})$ , pearlite. In the normal condition this 1.0 per cent carbon steel consists of less than 5 per cent "free" cementite crystals and over 95 per cent pearlite.

**270.** The above examples have shown how the normal structures originate and how the amounts of  $\alpha$  and  $(\alpha + \text{Fe}_3\text{C})$ , or  $(\alpha + \text{Fe}_3\text{C})$  and  $\text{Fe}_3\text{C}$ , change with the carbon content of the steel. The relative importance of the different structural components in a steel may be calculated from the inverse proportions of the distances of its composition from the composition of pure structures. A normal steel with zero carbon is 100 per cent ferrite. A normal steel with 0.9 per cent carbon is 100 per cent pearlite. A normal steel with 0.4 per cent carbon will consist  $\frac{0.4}{0.9}$  parts pearlite and  $\frac{(0.9 - 0.4)}{0.9}$  parts ferrite, or 44 per cent

pearlite and 56 per cent ferrite. The composition of pure  $\text{Fe}_3\text{C}$  is 6.7 per cent carbon. Hence a normal steel of 1.0 per cent carbon has an expected structure consisting of  $\frac{(6.7 - 1.0)}{(6.7 - 0.9)}$  parts, or 98 per cent, pearlite, and  $\frac{(1.0 - 0.9)}{(6.7 - 0.9)}$  parts, or 2 per cent, free cementite. The same method may be applied anywhere on the equilibrium diagrams to find the proportions of two phases or structures in the metal. The 0.2 per cent carbon steel at  $770^\circ \text{C}$ . is a mixture of  $\beta$  crystals of composition  $L$  and  $\gamma(\text{C})$  crystals of composition  $N$ , Fig. 94. As 0.2 carbon comes at  $\frac{1}{3}$  the way from  $L$  to  $N$  the metal will be at this temperature  $\frac{2}{3} \beta$  and  $\frac{1}{3} \gamma(\text{C})$ .

271. The engineering properties of the normal steels will depend upon the proportions of ferrite and pearlite or pearlite and cementite in the structure. These proportions are determined by the carbon content. The variation of the proportions with carbon content is summarized in Fig. 95. A vertical line  $CA$  across this figure, at any carbon content such as  $C$ , will be divided into portions  $CB$  and  $BA$ , proportionate to the amounts of pearlite and free ferrite or pearlite and free cementite in that steel. It is found that the amount of pearlite in steels exceeds the theoretical proportion to some extent, as indicated in the figure. Fig. 96 illustrates in another way the change of normal structures with change of carbon content of the steels.

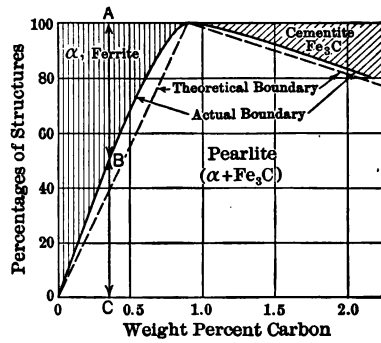


FIG. 95.

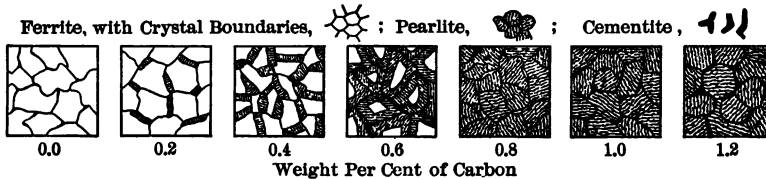


FIG. 96.

Fig. 96 is a conventionalized representation of the appearance of the different steels under the microscope.

**272.** The properties of the principal structural types in normal carbon steels are as follows: (1) Free ferrite,  $\alpha$ , is soft, weak, very ductile, strongly magnetic, of low yield point, and of low shear breaking strength. Its true tensile strength, if it could be measured, would probably be very high. The engineering tension yield point of ferrite is around 25,000 pounds per square inch and engineering maximum tension strength around 40,000 pounds per square inch. The shear yield point of ferrite is about 12,000 pounds per square inch and the shear breaking strength somewhat over 50,000 pounds per square inch. (2) Cementite,  $\text{Fe}_3\text{C}$ , is harder than glass, brittle, of very high shear breaking strength (probably above 300,000 pounds per square inch), but of low tensile strength. When free cementite appears in the normal steels (carbon content above 1 per cent) it tends to occur in the form of thin envelopes around patches of pearlite. Such a steel will break readily through the brittle cementite envelopes and will have its properties largely determined by the cementite, even though the amount of free cementite is quite small. By proper handling of a high-carbon steel the formation of the envelopes of cementite may be prevented. (3) Pearlite,  $(\alpha + \text{Fe}_3\text{C})$ , is a mechanical mixture in definite proportions of minute ferrite and cementite crystals. The crystals of ferrite and cementite are arranged in alternate layers. The individual crystals are of plate form, of the general dimensions about 0.001 inch long and broad, and 0.0001 inch thick. The amounts of ferrite and cementite in pearlite are 85 per cent ferrite and 15 per cent cementite, in parts by weight. The volume proportions are about 80 per cent and 20 per cent. From these volume proportions, it is evident that the hard cementite crystals may be considered as bedded in a soft matrix of ferrite. Pearlite has a tension yield point about 75,000 pounds per square inch and an engineering tension strength from 120,000 to 160,000 pounds per square inch. Both yield point and maximum strength are considerably modified by crystal size. When the ferrite and cementite crystals are very small the structure is called sorbite instead of pearlite and the engineering properties are better than for ordinary pearlite. The hardness of pearlite is about 2.5 times that of the ferrite. Pearlite is not hard enough to make a cutting edge for tools, but it is hard enough to require excellent tools to cut the pearlite.

**273.** The changes of engineering properties of normal steels with change of carbon content are practically in arithmetical proportion to the amounts of pearlite and ferrite (or cementite) present and the properties of these structural constituents. The forms of the curves of engineering properties (tension) versus carbon content are shown in Fig. 97. By comparison of data from French, English and American sources, it seems that up to one per cent carbon the minimum values of the various properties may be given by the following equations:

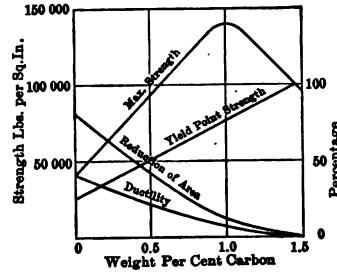


FIG. 97.

Engineering maximum tension strength, pounds per square inch =  $40,000 + 100,000 C$ , when  $C$  stands for the per cent of carbon.

( $C = 1.0$  for 1.0 per cent of carbon.)

Yield-point strength =  $25,000 + 50,000 C$ .

Ductility on a specimen 2 inches long by  $\frac{1}{2}$  inch diameter =  $50 - 48 C$ .

Ductility on a specimen 8 inches long by 0.8 inch diameter =  $32 - 30 C$ . (The form of the test piece has so large an effect on ductility that it is practically impossible to give equations for the effect of carbon content on ductility.) It may be noted that between 0.2 and 0.5 per cent carbon, crucible and basic open-hearth steels are slightly weaker than the above formula for maximum strength would indicate.

To understand definitely the relation of properties of normal steels to carbon content, a broader and deeper study is necessary. Fig. 98 outlines the relations. The curve  $AEG$  is the tension yield point stress  $p_E$ , which is twice the shearing yield point stress of the material. The curve  $BKE$  is the engineering nominal maximum stress  $p_{NM}$ .  $CLEH$  is the physical stress intensity  $p_B$  at the actual breaking load.  $CLF$  is about twice the shear breaking strength of the material or  $2 q_B$ .  $DEH$  is the curve of tension strength  $p_{TB}$  resisting pure tension break. These various curves cut each other off so that the dotted portions  $DL$ ,  $LF$  and  $EG$  are imaginary. With any carbon content less than about 0.7 per cent the steel will have a tension yield point  $p_E = 2 q_E$ . Beyond the yield point it will stretch rapidly,

with considerable permanent deformation, passing through a maximum load and breaking at a load lower than the maximum. The break will have a cup and cone form. (See sections 67, 68 and 69, and Fig. 23.) The break form will be practically all cone at zero carbon with

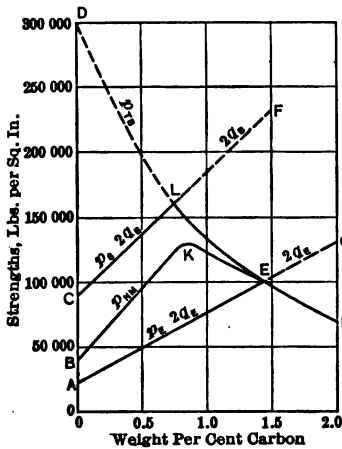


FIG. 98.

increasing flat part until the cone vanishes at  $0.7 \pm$  per cent carbon. The beginning of break is conditioned by the shear breaking strength of the material. Dividing actual load by actual area at break,  $\phi_B$  is approximately  $2 \phi_{NM}$ . From  $0.7 \pm$  per cent carbon to  $1.5 \pm$  per cent the yield-point action is similar to that in the lower carbon steels. The break is distinctly different, being a pure tension break, square across the piece. There is a little permanent deformation of the piece before break. The maximum load and breaking load of the

tension test are near together or are identical.  $\phi_B$  remains greater than  $\phi_{NM}$  because there is still some reduction of area before break. At the point E the yield-point curve and the curve of pure tension breaking strength cross. Beyond the point E, with higher carbons than 1.5 per cent, the normal steels will have no yield point but will break at maximum load, as brittle materials, with no appreciable permanent deformation. From zero to 0.7 per cent carbon the normal steels may be classed as fully ductile, from 0.7 per cent to 1.5 per cent as semi-ductile and above 1.5 per cent as brittle.

**274.** Besides carbon and iron in the ordinary steels there are several other chemical elements present in amounts not quite negligible. In the above discussions of the properties of the steel, it has been considered that these other elements were present in amounts characteristic of present commercial practice. The pure iron-carbon alloys would not have quite the properties outlined above for steel. The secondary elements can be divided into two groups. The first group is of injurious elements which we would eliminate if we could. Some of these elements are in the original ores and cannot be entirely re-

moved. Others are unavoidably introduced in the metallurgical processes. This group of injurious elements consists of sulphur, oxygen, arsenic, phosphorus and nitrogen. The second group consists of elements added to modify the bad effects of the injurious elements. Manganese and silicon are used to deoxidize the liquid steel, and the manganese puts the sulphur in a comparatively harmless condition.

The specific effects of these secondary elements are as follows:

Sulphur, present as MnS, slightly hardens and strengthens the steel in its normal condition. Sulphur is injurious because it causes hot shortness (see section 247). On this account the sulphur must be kept under 0.1 per cent.

Oxygen, probably as FeO, acts like MnS. Oxygen as SiO<sub>2</sub> is present in Bessemer steels in microscopic specks mechanically included in the ferrite crystals. It makes the Bessemer steels less ductile than corresponding open-hearth steels.

Arsenic acts like sulphur, but is usually negligible.

Phosphorus and nitrogen in small amounts, as they usually are (less than 0.1 per cent), are probably in solid solution in the ferrite. The presence of carbon lessens the amounts of phosphorus and nitrogen which can be held in solid solution. Free Fe<sub>3</sub>P or Fe<sub>2</sub>N probably form brittle envelopes on pearlite patches and are very dangerous. Above the critical temperatures P and N are in solid solution in the austenite and have no bad effects. At ordinary temperatures even slight excesses of P or N make the steel brittle against shock and this brittleness rapidly increases as the temperature falls. Present commercial practice specifies the amount of P in steels, but usually does not specify the limits for N. Probably the N is the more dangerous of the two.

Manganese and silicon are purposely put into the liquid steel at the end of the purification process. The purification is essentially an oxidation of all elements out of the liquid metal to the point of oxidation of the iron itself. The melt is deoxidized and brought to the composition of steel desired by the addition of special alloys containing usually manganese, silicon and carbon.

The amount of manganese in ordinary steels runs from 0.4 per cent to 1.0 per cent. In general the higher the carbon the lower the manganese; the decrease of manganese is approximately one-half the

increase of carbon. Manganese makes the steel harder and more brittle. The manganese is partly in solid solution in the ferrite, partly combined with sulphur as  $MnS$ , and partly combined with carbon as  $Mn_3C$ , which goes into solid solution in  $Fe_3C$ . In general the effect of manganese on the strength of normal steel is about one-third as great as that of the same amount of carbon.

Silicon is present in steels in amounts from 0.1 per cent to 0.3 per cent. Its effects are slight, being equivalent to an amount of carbon about one-fifth as great. In these slight amounts, silicon is always in solid solution in the ferrite.

In American iron ores there are usually slight amounts of copper. In the finished steel the copper is in solid solution in the ferrite. It used to be considered that copper caused hot shortness like sulphur, but this has been disproven. Probably the only effect of copper is the beneficial one of reducing the tendency of steel to corrosion. It seems possible that by leaving out manganese and adding copper, steels might be made of high resistance to corrosion; but this is not yet proven.

When it is not desired to carbonize the steel at the end of the purification process, manganese and silicon are not available as deoxidizers as in the ordinary commercial process. Aluminum may be used instead as a deoxidizer. The  $Al_2O_3$  formed floats to the top of the steel and is skimmed off as a slag.

Two purifying elements that may occasionally be used are titanium and vanadium. They are more effective than Mn and Si, but are too expensive for ordinary use. V is a very effective deoxidizer. V in slight excess of that required for deoxidation goes into solid solution in the ferrite with beneficial results. Ti is a good deoxidizer and in addition does what few other elements can do. It combines with N as  $Ti_3N_4$ , putting the N into a harmless condition. Ti seems also to have a peculiar effect of reducing segregation.



## CHAPTER XV

### GENERAL THEORY OF HEAT TREATMENT

275. The mechanical methods, hot or cold working, for controlling the properties of metals cannot be used to produce really radical changes in the properties. Great changes of properties, changes as great as those involved in the hardening of tool steel, are possible only by **heat treatment**. Heat treatment is, in general, a practical utilization of three facts: first, that many metals have at high temperatures a different internal structure from that at low temperatures; second, that change of crystal kind or size in the metals requires an appreciable time for completion at any temperature; third, that the time required for these changes, while reasonably small at high temperatures, approaches infinity at ordinary temperatures. Heat treatment controls the kind and size of the internal structure of the metals, and consequently their engineering properties. The discussion given here applies especially to "carbon" steels, but the general principles are the same for all metals.

By heating a "carbon" tool steel to a temperature above the critical, it changes spontaneously from the normal structure of  $\alpha$  and  $\text{Fe}_3\text{C}$ , ferrite and cementite crystals, to  $\gamma$  (C), austenite. (See Fig. 79.) If this steel were cooled slowly it would change back at the critical temperature to a mixture of  $\alpha$  and  $\text{Fe}_3\text{C}$ . The time required at red heat for this change, after it starts, is from ten to thirty seconds. If the steel is slowly cooled the change is completed while the temperature is still high, and at the end of the cooling the steel will be soft. By taking the hot steel from the furnace and plunging the piece into water, it is possible to cool the steel to below a red heat in three or four seconds. By such "quenching" we deny the steel time to complete the change from  $\gamma$  (C) to  $\alpha + \text{Fe}_3\text{C}$ . The quenched piece will be found to be very hard, and also brittle. At ordinary temperatures this hard brittle structure is practically permanent. If the hardened steel is warmed up, even to so low a temperature as that of boiling water, it begins to soften. The higher the reheating temperature the more rapid is this softening. As it softens, the hardened steel becomes

less brittle or tougher. Because the fully hardened tool steel is too brittle to be used, we commonly apply this softening and toughening by reheating, calling it "**tempering.**" By a proper amount of tempering we can make the hardened steel tough enough to use in tools while still retaining a large part of its full hardness.

**276.** Closer study of the hardening of steel discovers an apparent mystery. Austenite,  $\gamma$  (C), is soft and tough at any temperature. The ferrite and cementite mixture,  $\alpha + \text{Fe}_3\text{C}$ , is also comparatively soft. The hardened steel is two or three times as hard as either  $\gamma$  (C) or  $\alpha + \text{Fe}_3\text{C}$ . The quenching prevented the completion of the change from  $\gamma$  (C) to  $\alpha + \text{Fe}_3\text{C}$ . It is evident that this change cannot be a simple one. **From the microscopic examination and the accompanying measurement of various properties of the steel, we have found that the change from  $\gamma$  (C) to  $\alpha + \text{Fe}_3\text{C}$  has four stages. These are technically known as the austenitic, martensitic, troostitic and sorbitic conditions of the steel.**

The austenitic and the sorbitic stages are the chemically stable forms  $\gamma$  (C) and  $\alpha + \text{Fe}_3\text{C}$ .  $\gamma$  (C) is stable above and unstable below the critical temperature.  $\alpha + \text{Fe}_3\text{C}$  builds structures stable below and unstable above the critical temperature. The martensitic and troostitic stages are not stable under any conditions, from a theoretical standpoint, but if we can obtain them at ordinary temperatures they are practically permanent. **The martensitic stage is the hard brittle stage.** The hardness utilized in cutting tools comes from the martensite in the hardened steel. Troostite is similar in engineering properties to sorbite and pearlite. It is stronger and less ductile than sorbite, which in turn is stronger than pearlite.

When, by quenching, we prevent the  $\gamma$  (C) from completing the change to  $\alpha + \text{Fe}_3\text{C}$ , we usually catch a steel in the second stage of the change, the martensitic stage. Tempering changes the martensite to troostite or sorbite.

**277.** The internal changes in metals, the changes with which we deal in heat treatment, are chemical reactions. Our heat treatment consists in controlling these reactions, stopping them from going through to completion. It is necessary therefore to know how a chemical reaction goes on with time at constant temperature and how change of temperature changes the time rate of reactions.

A chemical reaction is not instantaneous. The reactions concerned in heat treatment consist usually in the decomposition of a single substance into one or more different substances, or the reverse action.

When a single substance decomposes, the mathematical statement of the law which has been experimentally proved for such cases is that a certain fraction of the number of molecules present at the beginning of a given unit of time will decompose during that unit of time. For example, suppose  $\gamma$  iron is decomposing to form  $\beta$ .<sup>\*</sup> The curve showing the amount of  $\gamma$  remaining at any instant of time in the reaction is shown in Fig. 99. The material which is no longer  $\gamma$  is of course now  $\beta$ . Taking the unit of time as that time in which the first 50 per cent of the  $\gamma$  disappears, the amount of the  $\gamma$  remaining after ten units of time is 0.1 per cent and after twenty units of time is 0.0001 per cent. Perhaps it might be said in

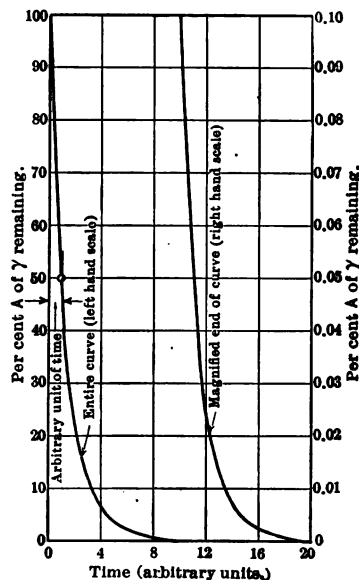


FIG. 99.

a mathematical sense that the reaction is never quite complete; but completion to the degree 0.999 is reached in ten of the above units of time, and to the degree 0.99999 in twenty time units. For practical purposes then the reaction would be complete in about ten such time units.

**278.** The decomposition of  $\gamma$  (C), in its change to  $\alpha + \text{Fe}_3\text{C}$ , goes through four successive stages, austenite, martensite, troostite and sorbite. Let the percentages of these four stages at a given time be

<sup>\*</sup> Let  $A$  stand for the number of  $\gamma$  molecules present at a given instant. The mathematical statement becomes  $\frac{1}{A} \cdot \frac{dA}{dt} = -K$ . ( $t$  stands for time and  $K$  is a constant.) Integrating this differential equation,  $\log_e A = M - Kt$ , or  $A = e^{(M - Kt)}$ . The constant of integration,  $M$ , is determined by the amount of  $A$  present at zero time or at the beginning of the reaction. Stating the amount of  $A$  in percentage, with the initial percentage of  $A$  equal to 100, the useful form of the equation is  $\log_{10} A = 2 - Kt$ , or  $A = 10^{(2 - Kt)}$ .  $K$  measures the time rate of the reaction.

*A*, *B*, *C* and *D* respectively. The general form of the mathematical expectation of the variation of *A*, *B*, *C* and *D* with time is shown in Fig. 100, and is also tabulated herewith.

Units of time (arbitrary)	Percentage each structure is of total				Actual time scale for 0.7 % carbon steel, at 680° C., temperature of maximum rate of reaction
	<i>A</i>	<i>B</i>	<i>C</i>	<i>D</i>	
0	100.0	0	0	0	0
1	60.0	40.0	0	0	0.2
2	36.0	56.0	8.0	0	0.4
3	21.6	59.2	18.4	0.8	0.6
4	13.0	56.0	28.4	2.6	0.8
5	7.8	50.0	36.8	5.4	1.0
6	4.7	43.1	43.1	9.1	1.2
7	2.8	36.4	47.4	13.4	1.4
8	1.7	30.2	50.0	18.1	1.6
9	1.0	24.9	51.0	23.1	1.8
10	0.6	20.3	50.9	28.2	2.0
12	0.2	13.3	48.2	38.3	2.4
14	0.1	8.6	43.6	47.7	2.8
16	0.0	5.6	38.2	56.2	3.2
18	0.0	3.6	32.8	63.6	3.6
20	0.0	2.3	27.8	69.9	4.0
22	0.0	1.4	23.3	75.3	4.4
24	0.0	0.9	19.4	79.7	4.8
26	0.0	0.6	16.0	83.4	5.2
28	0.0	0.4	13.1	86.5	5.6
30	0.0	0.2	10.8	89.0	6.0
35	0.0	0.1	6.4	93.5	7.0
40	0.0	0.0	3.9	96.1	8.0
45	0.0	0.0	2.0	98.0	9.0
50	0.0	0.0	1.0	99.0	10.0
55	0.0	0.0	0.5	99.5	11.0
60	0.0	0.0	0.3	99.7	12.0
70	0.0	0.0	0.2	99.8	14.0
80	0.0	0.0	0.2	99.8	16.0
90	0.0	0.0	0.1	99.9	18.0
100	0.0	0.0	0.0	100.0	20.0

The specific assumptions in calculating the quantities for this table and figure were that the fraction of *A* changing to *B* per unit of the arbitrary time is 0.4 of the *A* present; that, of the *B* present at the beginning of a given unit of time, 0.2 changes to *C* during that unit of time; while the *C* to *D* change has a constant of 0.1 per unit of time. During a given unit of time the amount of *B* is increased by decomposition of *A*, but is also decreased by change of *B* to *C*. The amount of *B* present at any given instant is the summation of the results of these

two actions from the beginning of the complex change. No matter what the assumptions of constants for the successive changes, curves of the same general type as in Fig. 100 will always be found. Change in the reaction rate constants affects only the positions and heights of the maxima of the *B* and *C* curves, and the final time for the completion of the whole reaction. Changes in the rate constants occur from the change of the chemical composition of the steel and from change of temperature. Fig. 100 is perhaps relatively correct for a carbon steel

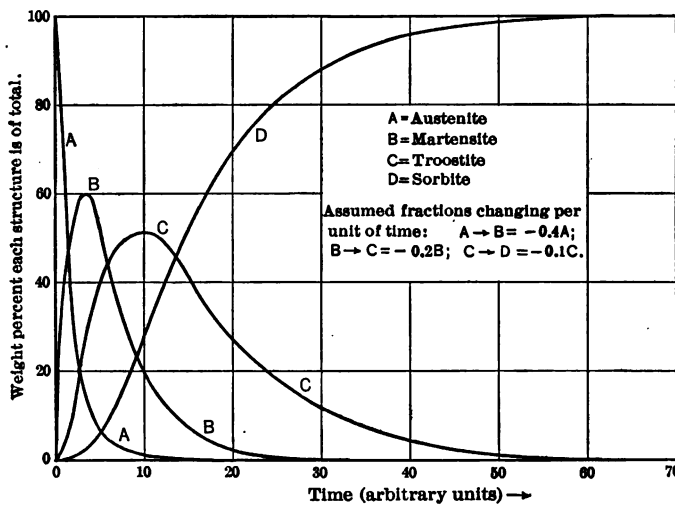


FIG. 100.

of 0.7 per cent carbon. (See also Figs. 103, 105, 106 and 107, which have the time scale in seconds.)

**279.** Fig. 100 gives a new understanding of the structures of heat-treated steels. The only forms which can constitute 100 per cent of the structure are the two stable end forms of the reactions, austenite and sorbite. A steel cannot be entirely martensite or entirely troostite. A martensitic steel is dominantly martensite, but contains also austenite and troostite in considerable amounts. Similarly a troostitic steel contains appreciable amounts of martensite and sorbite.

**280.** Change of temperature has a very great effect upon the time rate of chemical reactions. A common statement is that the rates of action double for every rise of  $10^{\circ}$  C. at or near ordinary temperatures.

The effect of temperature upon the time rate of reaction is a specific effect for each reaction.

The reaction  $\gamma(C) \rightleftharpoons \alpha + Fe_3C$  is a reversible reaction, proceeding to the left if the temperature is above the critical ( $720^\circ C.$  in carbon steels) and to the right if the temperature is below the critical. At the critical temperature itself  $\gamma(C)$  and  $\alpha + Fe_3C$  can coexist indefinitely, in analogy with water and ice at the freezing point. There are several factors in the general equation for reaction rate or reaction time as a function of temperature. (1) The releasing factor,  $(T_c - T)$ , difference of temperature from the critical temperature  $T_c$ . The tendency of change from  $\gamma(C)$  to  $\alpha + Fe_3C$ , for instance, must increase as the temperature falls away from the critical temperature where equilibrium would exist. (2) Energy released by the change. This is, at the critical temperature, the latent heat of the change. The latent heat of change varies with the temperature on account of the difference in the specific heats of the phases or sets of phases. For example, the latent heat of change of water to ice is 80 calories per gram at  $0^\circ C.$ , about 75 calories at  $-10^\circ$ , 70 calories at  $-20^\circ$ , etc. It is perhaps true that the smaller the latent heat, the more rapid the reaction. (3) The brake factor,  $\phi(T)$ , an unknown function of the absolute temperature. The brake factor is probably a complex of two terms, one a specific term for the particular reaction in question and the other a general term showing the effect of temperature common to all reactions. It will answer our purpose, which is merely that of sketching out the problem, to ignore the change of latent heat with temperature, and to consider  $\phi(T)$  an exponential function of  $T$ , or  $T^n$ . The exponent  $n$  may be approximated from two known facts. A doubling of reaction rate at ordinary temperatures, for a rise of  $10^\circ C.$  in temperature, indicates that  $n =$  about 20. The softening of hardened steel takes a time of the order of a thousand years at ordinary temperatures, and only a few seconds at  $700^\circ C.$  This likewise indicates  $n = 20$ . We may then write the equations:

$$\begin{aligned}\text{Rate of action} &= K_1 (T_c - T) T^{20}, \text{ and} \\ \text{Time for reaction} &= K_2 (T_c - T)^{-1} T^{-20}.\end{aligned}$$

$K_1$  and  $K_2$  are constants and  $T$  signifies absolute temperature.

In the above equations the term  $(T_c - T)$  makes the rate of action

zero, and the time for completion of reaction infinity, when  $T = T_c$ , as should be the case. The same term reverses the direction of the reaction as  $T$  passes  $T_c$ , which is again as it should be. The equation assumes that the reaction begins in one direction or the other as soon as the temperatures leave the critical temperature. The suspension of change during superheating or supercooling is not represented by the equations; nor are pressure effects accounted for. Fig. 101 shows the form of the calculated rate of reaction from the equation; the numerical values are given in the table on page 178.

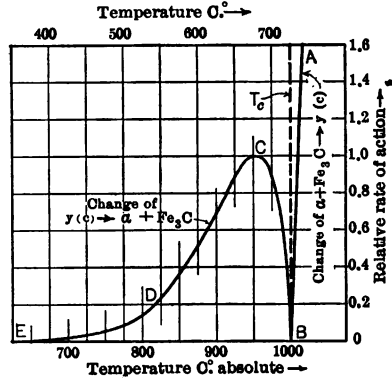


FIG. 101.

The calculations were made for steel, assuming the critical temperature at  $727^\circ\text{C.}$ , for convenience, because this makes  $T_c$  equal to  $1000^\circ\text{C. absolute}$ . The constant  $K_1$  was fixed by making the maximum rate of action, below the critical temperature, at the point C in Fig. 101, equal to unity. Along the branch BA of the curve, Fig. 101, both change of the  $(T_c - T)$  term and increase of absolute temperature work together to make the reaction rate increase rapidly as the temperature goes above the critical. This corresponds to the experimental fact that it is impossible to superheat a solid phase to any considerable degree. Along the branch BCDE of the curve, below the critical temperature, increase of  $(T_c - T)$  at first dominates, increasing the reaction rate from B to C. The  $(T_c - T)$  term is opposed by the change of  $T^n$ . The factor  $T^n$  dominates the form of the curve along CDE. The curve BCDE corresponds quite well to the known action of steel in heat treatment. There is a region between  $700^\circ$  and  $500^\circ\text{C.}$  where changes in steel during heat treatment are quite rapid. This region we often speak of as the "critical range" of temperatures in heat treatment of steel, because in order to harden steel we must cool rapidly through all of this range of temperature. In the tempering of tool steel experience indicates that changes begin to be quite rapid between  $400^\circ$  and  $500^\circ\text{C.}$  There is no intention of giving the

above equations or the curves of Fig. 101 as quantitatively correct or precise, but it is evident that they are qualitatively of the right form.

RELATIVE RATE OF REACTION IN INTERNAL CHANGES OF STEELS, AS A FUNCTION OF TEMPERATURE (EXCLUDING PRESSURE EFFECTS, SUPERCOOLING AND SUPERHEATING)

Absolute temperature, °C.	Temperature, C.	Temperature difference from the critical, °C.	Relative rate of reaction	
1100	827	100	37.5	Change of $\alpha + \text{Fe}_3\text{C}$ toward $\gamma$ (C).
1080	807	80	20.8	
1060	787	60	11.0	
1040	767	40	4.88	
1030	757	30	3.02	
1020	747	20	1.66	
1010	737	10	0.680	
1000	727	0	0	
990	717	10	0.456	Change of $\gamma$ (C) toward $\alpha + \text{Fe}_3\text{C}$ .
980	707	20	0.744	
970	697	30	0.909	
960	687	40	0.985	
950	677	50	0.999	
940	667	60	0.970	
930	657	70	0.913	
920	647	80	0.841	
910	637	90	0.778	
900	627	100	0.677	
850	577	150	0.324	
800	527	200	0.129	
750	477	250	0.0442	
700	427	300	0.0134	
650	377	350	0.00354	
600	327	400	0.000815	
550	277	450	0.000161	
500	227	500	0.0000266	
450	177	550	0.00000355	
400	127	600	0.000000368	
350	77	650	0.0000000276	
300	27	700	0.00000000136	
250	-23	750	0.00000000038	

281. When the calculation of reaction time is made over a wide range of temperature, the range of time values is such that the time cannot be shown on a curve except by the use of logarithmic scaling. Fig. 102 is a plot of the equation (time for reaction) =  $K_2 (T_c - T)^{-1} T^{-20}$ , with logarithmic scaling for the ordinate, the reaction time.  $K_2$  is determined by making the minimum time for the reaction, below the critical temperature, equal to unity. This minimum time occurs at about 680° C., according to the equation. This



again checks with experience that the supercooling of the high carbon steels, in air cooling, usually terminates between  $700^{\circ}\text{C}.$  and  $650^{\circ}\text{C}.$  The information of Fig. 102 is perhaps more interesting when pre-

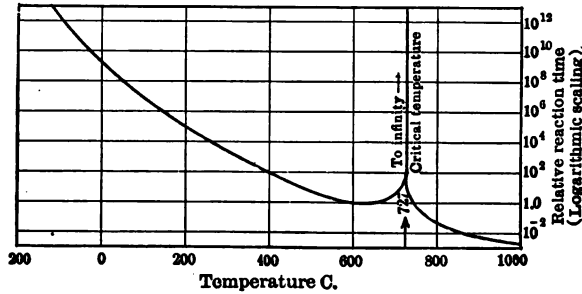


FIG. 102.

sented in the form of a table. As a basis for the calculations, the assumption was made that the unit time for completion of reaction, Fig. 102, is one minute. We know by experience that this is approximately correct for steel. The table of calculated reaction time follows:

Reaction	Temp. $^{\circ}\text{C}.$	Time for completion of reaction
$\alpha + \text{Fe}_3\text{C} \rightarrow \gamma (\text{C})$	900	0.6 seconds
	775	6.0 seconds
	727	Infinity
	680	1 minute
	540	6 minutes
$\gamma (\text{C}) \rightarrow \alpha + \text{Fe}_3\text{C}$	430	60 minutes
	370	6 hours
	210	60 days
	125	6 years
	80	60 years
	40	600 years
	0	6000 years
Equilibrium . .		

The times required for softening an already hardened steel are perhaps one-fifth to one-half of the times for the complete reaction  $\gamma (\text{C})$  to  $\alpha + \text{Fe}_3\text{C}$  in the table above. Both this statement and the table itself assume no internal pressure effects on the rate of action. Such pressure effects always do occur in the heating or cooling of steel and have considerable effects on the results of heat treatments.

**282.** The curves of progress with time of the *A-M-T-S* (austenite

→ martensite → troostite → sorbite) transformation were given in Fig. 100 with arbitrary time units. They depict the change in a carbon steel of about 0.7 per cent carbon. Fig. 103 repeats these curves, now with a scaling of actual time in seconds, for the suppositious case that the reaction starts and proceeds to completion at uniform rate, at the temperature of  $680^{\circ}\text{C}$ . In Fig. 101 it will be seen that this temperature gives the maximum, and what we have taken as the unit, rate of the *A-M-T-S* change.

Actually the transformation does not, and cannot, take place as shown in Fig. 103. In the actual case the transformation is due to begin as soon as the falling temperature of the metal passes the critical

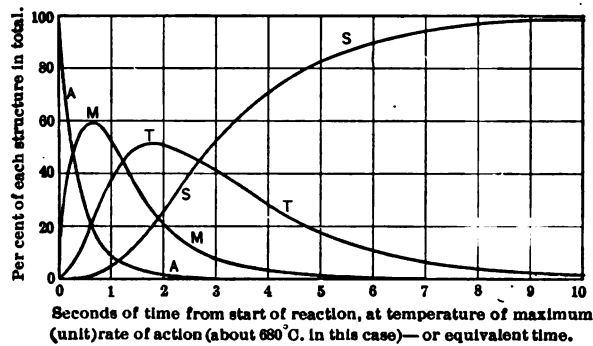


FIG. 103.

temperature. As the temperature continues to fall the rate of the internal changes will vary with temperature as shown in Fig. 101. At the critical temperature the rate will be zero. The rate rapidly increases until the temperature is about  $50^{\circ}\text{C}$ . below the critical (point C, Fig. 101); then almost as rapidly decreases as the temperature goes still lower (*CDE* of curve, Fig. 101).

283. To find, from Fig. 103 as base, what the transformation curves are when the temperature is varying during the transformation, we make use of the conception of "equivalent time." "Equivalent time" is the product of actual time spent at a given temperature by the relative rate of action at that temperature. If temperature is changing, equivalent time must be found by an integration. The table below gives the method of calculating equivalent times from actual times, during cooling of the metal, when cooling takes place at a uni-

form rate. The last column of the table gives the factors for reducing actual time to equivalent time, for this special case.

TABLE DERIVING THE FACTORS FOR REDUCING ACTUAL TIME, AFTER PASSING OF THE CRITICAL TEMPERATURE, TO EQUIVALENT TIME, IN COOLING AT UNIFORM RATE

Degrees below critical temperature	Relative rate of action, at end of interval, = $r$ .	Temperature interval = $\Delta T$	Relative increment of action = $r \Delta T$ , the value of $r$ being the average during the interval	Accumulated amount of relative action = $\int r \cdot \Delta T$	Ratio of accumulated action to total degrees below critical temperature or average relative rate of action over the temperature range from the critical temperature.
0	0	0	0	0	0
2	0.107	2	0.107	0.11	0.0536
4	0.206	2	0.313	0.42	0.1050
6	0.296	2	0.502	0.92	0.1537
8	0.379	2	0.676	1.60	0.1997
10	0.456	2	0.835	2.43	0.2432
15	0.618	5	2.683	5.11	0.3410
20	0.744	5	3.404	8.52	0.4259
25	0.839	5	3.961	12.48	0.4992
30	0.909	5	4.371	16.85	0.5617
40	0.985	10	9.468	26.32	0.6580
50	0.999	10	9.618	35.94	0.7187
60	0.970	10	9.844	45.78	0.7630
70	0.913	10	9.417	55.20	0.7885
80	0.841	10	8.773	63.97	0.7996
90	0.778	10	8.100	72.07	0.8008
100	0.677	10	7.277	79.25	0.7925
150	0.324	50	25.04	104.28	0.6952
200	0.129	50	13.81	118.09	0.5905
250	0.0442	50	4.32	122.41	0.4896
300	0.0134	50	1.44	123.85	0.4128
350	0.0035	50	0.42	124.27	0.3550
400	0.0008	50	0.11	124.38	0.3109
450	0.0002	50	0.02	124.40	0.2764
500	0.0000	50	0.01	124.41	0.2488
600	.....	100	0.01	124.42	0.2074
700	.....	100	0.00	124.42	0.1777

Fig. 104 is a plot of the relation of equivalent time vs. actual time after passage of the critical temperature for the case of cooling at the rate of  $50^\circ$  per second. The abscissas of Fig. 104 are doubly scaled in seconds of actual time after passage of the critical temperature, and in degrees below the critical temperature. A series of points for the curve were calculated by means of the table above. For example, at actual time two seconds, with temperature attained  $100^\circ$  C. below

the critical, the table shows the average relative rate of action to that point to have been 0.7925. The equivalent time is then  $2 \times 0.7925 = 1.585$  seconds.

In the table of rate of action, and in Fig. 101, it will be noticed that

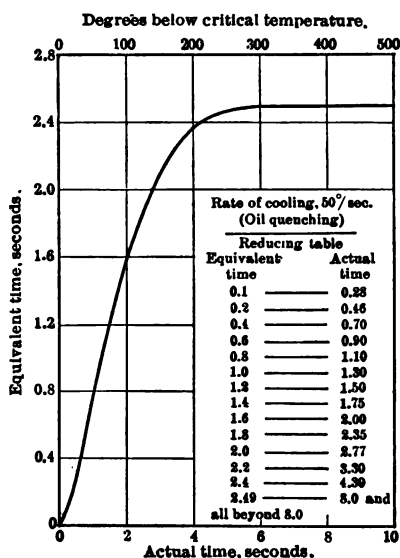


FIG. 104.

the rate of action becomes practically zero at about 300° C. below the critical temperature. As a result, any actual time at temperatures more than 300° C. below the critical causes practically no increase of equivalent time. This explains the flattening of the equivalent time curve of Fig. 104 after the actual time of six seconds.

**284.** The curves of transformation in Fig. 103 are really plotted to an abscissa of equivalent time units. A relation between equivalent time and actual time, such as has been worked out in Fig. 104, may now be used as an operator to put the transformation curves

on a basis of actual time. We have merely to pick off, for a sufficient number of points, the ordinates of the *A-M-T-S* curves in Fig. 103, and plot the same ordinates with new values of abscissas found from the relation between equivalent time and actual time.

Fig. 106 shows the effect of cooling at 50° per second, changing the curves of Fig. 103 from the basis of equivalent to that of actual time, by the procedure explained above. Like Fig. 104, Fig. 106 has two scalings of abscissas, one of time and one of temperature. Again like Fig. 104 no further appreciable changes occur after the temperature is more than 300° C. below the critical. **Fig. 106 shows that the cooling at 50° C. per second suppresses the final portion of the *A-M-T-S* change.** The quenched steel consists mostly of troostite and sorbite with an appreciable admixture of martensite.

Figs. 105, 107 and 108 were worked out in the same way as Fig. 106 for the rates of cooling in degrees per second of 1° C., 100° C. and

150° C. The four Figs. 105 to 108 cover the range of the commercial rates of cooling. At 1° C. per second, which corresponds to air cooling, it will be seen that the change to the final sorbitic condition is

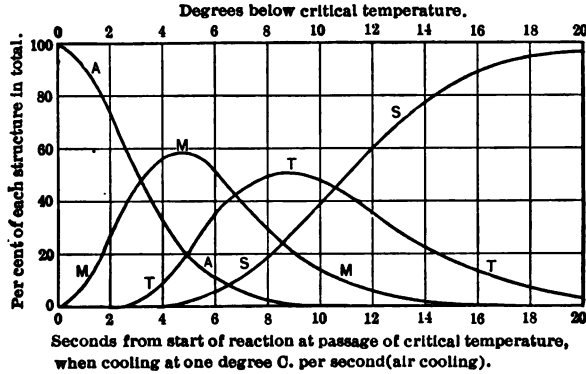


FIG. 105.

substantially complete at 25 or 30 seconds after passing the critical temperature, or within as many degrees of the critical temperature. The properties of the steel at the end of the air cooling are those we

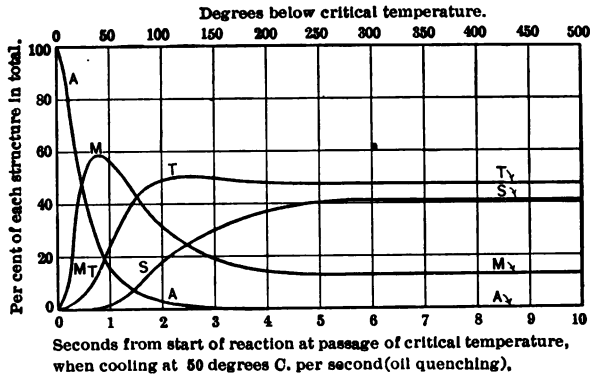


FIG. 106.

have called the "normal" properties of the soft steel. The steel will have reverted to the condition it had before the heating and cooling. Fig. 106 shows that 50° C. per second is a cooling sufficiently rapid to prevent the return to the normal structure and properties. The steel is appreciably hardened though probably not hard enough for cutting

tools. Cooling at  $100^{\circ}\text{C.}$  per second, represented in Fig. 107, causes still more suppression of the  $A-M-T-S$  transformation. The structures now dominating in the quenched steel are martensite and

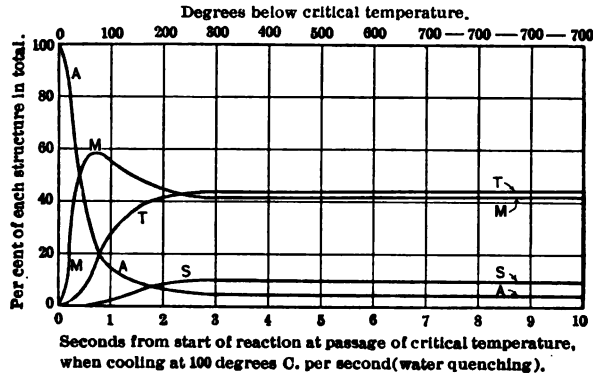


FIG. 107.

troostite. The steel will be very hard, and possibly brittle, on account of the large amount of martensite retained.

With commercial sizes of pieces it is practically impossible to reach rates of cooling during quenching much in excess of  $100^{\circ}\text{C.}$  per second.

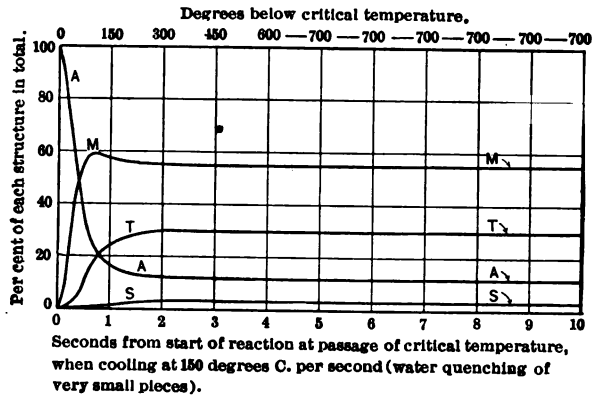


FIG. 108.

The effect of cooling at  $150^{\circ}\text{C.}$  per second has, however, been worked out, and is given in Fig. 108. The final result is a structure strongly martensitic, with minor amounts of troostite, austenite and sorbite. The hardness would not be greatly changed from that attained at

cooling at  $100^{\circ}\text{C.}$  per second, in this particular steel; but the brittleness is greatly increased, because there is no large amount of any ductile structure to make up for the brittleness of the martensite.

**285.** Comparing Figs. 105 to 108 with Fig. 103, it will be noticed that the point at which the  $A-M-T-S$  transformation is stopped by the cooling of the steel is less on the equivalent time scale of Fig. 103 when the rate of cooling is greater. In each separate case of cooling the equivalent time runs up to a limiting value, and there stops, as shown in Fig. 104. For the change of relative rate of action with temperature which we have assumed in Section 280 and Fig. 101, the limiting value of equivalent time for a given rate of cooling is precisely 124.4, or practically 125, divided by the rate of cooling in degrees C. per second. Thus we may tabulate:

Rate of cooling, degrees per second	Actual time, seconds, to cool from critical temperature to room temperature	Limit of equivalent time of the cooling, seconds, equals 124.4 divided by rate of cooling
1	700.0	124.4
10	70.0	12.44
50	14.0	2.49
100	7.0	1.24
150	4.7	0.83

This relation between the rate of cooling and the limiting value of the equivalent time for the cooling is independent of the composition of the steel as long as the critical temperature remains near  $700^{\circ}\text{C.}$ , as it is in the carbon steels. The relation may be applied in two ways. If the time scale of abscissas for the  $A-M-T-S$  transformation is known, as for instance it is given for a 0.7 per cent carbon steel in Fig. 103, the result of cooling at any given rate can immediately be read off. For example, cooling at  $100^{\circ}\text{C.}$  per second gives an equivalent time limit of 1.24 seconds. Entering Fig. 103 at the abscissa 1.24, the ordinates read off will be found to be the same as the final values for  $A, M, T$  and  $S$  in Fig. 107. We therefore have by this method a shortcut to the final results of any cooling. The second application of the relation between rate of cooling and limiting value of equivalent time is in establishing the time scale for the abscissas of such transformation curves as shown in Fig. 100. In establishing the time scale by

which Fig. 100 was converted into Fig. 103, it was known that for a 0.7 per cent carbon steel oil quenching ( $50^{\circ}\text{C.}$  per second cooling rate) gave martensite and troostite without enough martensite to make cutting hardness; water quenching of ordinary sized pieces ( $100^{\circ}\text{C.}$  per second) gave a strongly martensitic structure, brittle and hard; while water quenching of very small pieces was unable to catch any appreciable amount of austenite. Knowing the limiting equivalent time values for these different rates of cooling, and the structures obtained, it became evident that multiplying the arbitrary time scaling of Fig. 100 by 0.2 would reduce it to a scaling of equivalent seconds which would fit the case for the 0.7 per cent carbon steel. By similar trial with carbon steel of any other composition a time scaling for its transformation curves may be determined.

286. At several places in the preceding text statements of rate of reaction as controlled by the temperature have been qualified by the premise that pressure effects did not occur. Severe internal pressures always do occur in steel being heated or cooled, because of the temperature differences between outside and inside of the material and consequent differences of expansion. In the rapid cooling of steel the outside of the piece tries to shrink faster than the inside will allow, and so causes heavy compression stresses on the inner parts, and corresponding tension stresses in the outer parts of the material. These compressions or tensions would have no result if the densities of austenite, martensite, troostite and sorbite were the same. It happens, however, that martensite is distinctly greater in volume, or less in density, than the other two forms. Tensions, therefore, favor the formation of martensite, and compressions work against the formation of martensite, or for its disappearance if already present. At the start of a quenching in heat treatment the steel consists probably entirely of the comparatively dense austenite. The high internal pressure during the quenching hinders the transformation of austenite to martensite in the inside of the material, while simultaneously the tension in the outer surfaces favors the transformation. At the end of the cooling the temperatures of course equalize through the material. The internal stresses consequently reverse, because as the austenite core cools off it in turn shrinks compared to the already cold outside parts. The tendency then is to transform the core from austenite to



martensite at the end of the cooling, in contrast to the transformation of the outer shell at the beginning of the cooling. These pressure effects are superposed upon the general temperature effects. In very high carbon steels it has been found possible to keep the core in the austenitic condition, because the pressure effect at the end of the cooling was not strong enough to counteract the slowing up of the reaction by low temperature. A still more peculiar example of pressure effects has been found in some manganese-carbon steels which can be held almost entirely austenitic by quenching in water. The quenched pieces of these steels change from austenite to martensite by being plunged in liquid air and heated up to room temperature. This result is almost entirely an effect of shrinkage pressures, and the action takes place at the higher end of the temperature range. The outside of the piece becomes martensitic at the beginning of the cooling in liquid air, and the inside of the piece becomes martensitic during the reheating.

**287.** The preceding sections enable us to outline the nature of the phenomena of the hardening and tempering of tool steel. The tool steel is taken as a specific case of heat treatment. The general nature of the control of phases and structures is the same for almost any case of heat treatment of metals. Hardening, as in steels, is an exceptional phenomenon. The steel is first heated to a temperature somewhat above the critical and for a time long enough for the structure to become austenitic. (We must be careful not to heat the steel too long or too hot in the austenite field, for we must avoid crystal growth of the austenite, which makes the steel coarse grained and perhaps spoils it.) During the rapid cooling in quenching the steel is allowed to spend but a few seconds of time in the critical range of temperature,  $700^{\circ}$  to  $500^{\circ}$  C., where internal reactions go on rapidly. By this denial of time the *A-M-T-S* transformation is stopped considerably short of completion. In tool steel the reaction gets to the stage of dominant martensite, most of the austenite having disappeared, and considerable troostite having been formed. The steel is hardened because it contains considerable amounts of the extremely hard martensitic structure. After the quenching, theoretically, the reaction toward the final sorbitic condition goes on at ordinary temperatures. The rate of this action is, however, so very small that it would take several centuries for the hardened steel to become soft by secular

change at ordinary temperatures. (See table at end of Section 281.) The tool steel will be worn out in service long before its secular softening could become perceptible.

**288. Tempering** consists of raising the temperature of the hardened steel to such a point that the reaction toward the final soft condition begins to go on at a measurable rate, holding the steel at that temperature until the desired degree of softening is obtained, then rapidly cooling again to fix the desired condition. Fixation is possible because the rate of change at room temperature is negligible. For example, a steel when hardened might have reached the stage corresponding to abscissa 5 on the time scale of Fig. 100. By tempering we can let it go on at our pleasure as far as abscissa 10, or 15, or 20, etc., and then fix it by cooling. Since martensite, represented in amount by the *B* curve in Fig. 100, is the only extremely hard structure in the tool steel, this tempering process in general decreases the hardness of the steel, because it reduces the amount of martensite.

**289.** In practice the amount of change in the steel by tempering is regulated almost entirely by changing the temperature of the tempering, while the actual time employed remains small and nearly constant. The time scale of Fig. 100 corresponds roughly to seconds of actual time at about  $700^{\circ}\text{C}$ . The actual time for the same amount of change at any other temperature may be found by multiplying the abscissas of Fig. 100 by factors from the relative time scale of Fig. 102. For example, a change in steel which takes one minute at  $700^{\circ}\text{C}$ . will take at  $400^{\circ}\text{C}$ .  $10^2 = 100$  times as long of actual time, or approximately two hours. At  $200^{\circ}\text{C}$ . the factor from Fig. 102 is  $10^5$ , so that the change which takes one minute at  $700^{\circ}$  and two hours at  $400^{\circ}$  requires about 100 days at  $200^{\circ}\text{C}$ . These figures indicate the enormous effect of temperature upon the rate of reactions, and show one reason why control of temperature in tempering is the easiest and most important way of controlling the degree of tempering.

**290.** Pressure effects are of even greater commercial importance in the tempering than in the hardening of steel. The hardened steel, in the martensitic condition, is of lower density than the softer troostitic or sorbitic structures toward which the tempering is allowing the steel to go. It follows that compression stresses will accelerate the change of martensite to troostite, while tension stresses will delay the

change. Tempering is almost universally carried out by putting the piece of hardened steel into a furnace or bath which is already at the desired tempering temperature. The pieces of hardened steel are therefore rapidly heated; indeed the higher the tempering temperature, the more rapid is the initial rate of heating. In the rapid heating the outside of the steel is considerably hotter than the inside. Because of this difference of temperature and the tendency of the steel to expand with temperature rise, the inside of the piece experiences a heavy tension stress while the outside has corresponding compression. Therefore during the heating the outside of the piece changes from martensite to troostite much more rapidly than the expectation from the temperature alone. At the end of the heating the temperature equalizes throughout the piece. The outside of the piece, having already gone over to a condition of greater density, is now too small to suit the still martensitic core and there results a condition of compression stress in the core. This compression stress together with the high temperature causes an abnormally rapid change of martensite to troostite in the core.

If there were no pressure effects the progress of changes in tempering of hardened steel, as a function of time of heating, would be theo-

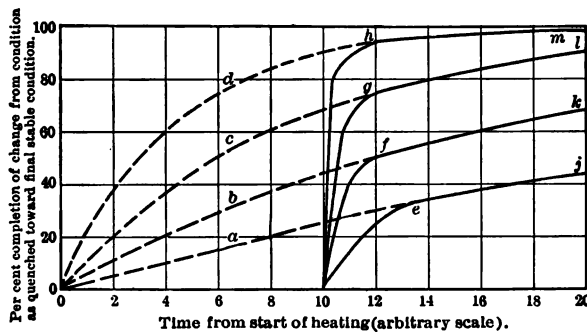


FIG. 109.

retically represented by the curves *oaej*, *obfk*, etc., in Fig. 109. The lower curves are for the lower tempering temperatures and vice versa. The addition of the pressure effect makes the actual curves of the results in the tempering take the form of *ioej*, *iofk*, etc., practically as if the origin of the time action had been shifted. The higher the

tempering temperature, the steeper are the actual curves from the origin at 10, because the pressure effect is greater in the more rapid heating.

Because the actual curves of results in tempering have the form of the full-lined curves in Fig. 109 from the origin at 10, due to the pressure effect, we have a second and very important reason for the choice of temperature in tempering, rather than time, for controlling the amount of change during tempering. If we heat the pieces in the tempering furnace or bath just long enough to be sure that they have come throughout to the desired temperature, we will have taken the pieces a bit beyond the bends *e*, *f*, *g* and *h* of the curves of Fig. 109. The portions *ej*, *fk*, etc., of the curves represent action at constant temperature without pressure effect; action so much slower than that during the heating of the pieces as to cause little additional result with time.

To show how the curves of Fig. 109 correspond to experimental results, the data of Barus and Strouhal is plotted in Fig. 110.\* The measurement used was that of thermo-electromotive force against silver. Change of this is probably due to the transformation of martensite to troostite more than to any other factor. The times of heating in this work go far beyond those used in commercial tempering operations.

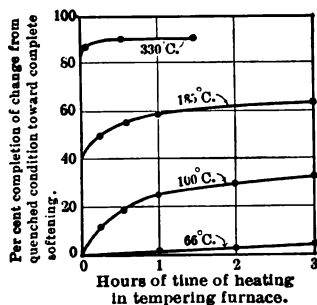


FIG. 110.

291. The most important specific factor in steels controlling the possibilities

in heat treatment is the carbon content of the steel. The higher the carbon content, the greater is the time required for the decomposition of  $\gamma$  (C) and the formation of the structure  $\alpha + \text{Fe}_3\text{C}$ . When no carbon is present, and we deal with nearly pure iron, it is almost impossible to accomplish anything by heat treatment. No matter how drastic the quenching, the iron will go during the time of quenching nearly to the final equilibrium condition. The abscissas 30 or 40 of the time scale of Fig. 100 will be passed. When the carbon content passes 0.5 per cent the changes in the structure of the steel

\* Taken from paper by Heyn and Bauer in *Iron and Steel Inst. Jour.*, 1909, I, page 125.

are slow enough so that the steel may be hardened sufficiently to make cutting tools. The higher the carbon content the greater is the amount of martensite which we can retain in the quenched steel. This does not mean that the peak of the *B* curve in Fig. 100 is higher as the carbon content of the steel increases. It means that with increase of carbon content the steel does not get so far, during quenching, along the descending branch of the *B* curve, beyond abscissa 5. The change of carbon content of the steel may be said to have the effect of altering the value of the units of the time scale of Fig. 100. When the carbon content of the steel is made as high as 1.5 per cent the changes are so slow that it is possible by quenching to stop the changes at a stage to the left of the peak of the *B* curve, Fig. 100. The steel may be made to bring appreciable amounts of austenite to ordinary temperature. **Increasing the carbon content of the steel used with a given rate of cooling in quenching has the same result in the final structure obtained as increasing the rate of cooling with a given carbon content.**

Of chemical elements other than carbon the only one of importance in the heat treatment of the carbon steels is manganese. Manganese is about as effective in slowing up the changes as would be an amount of carbon one-half as great in weight percentage as the manganese. Many other chemical elements have effects upon the heat treatment to be discussed in the chapter on alloy steels.

**292.** The most important scientific method of studying the changes of structure of steel in heat treatment is by microscopic examination of a polished and etched section through the steel. What the engineer cares about is the variation of the different properties of the steel as the structures change. An excellent set of data showing the changes in several physical properties during the progress of heat treatment is given in an article by E. Maurer in *Metallurgie* 6, 2, Jan., 1909.

From this material three sets of data have been selected. A steel with 1.94 per cent carbon and 2.2 per cent manganese, quenched in cold water from 1060° C., was found to be entirely (?) austenitic after quenching. A carbon steel of 1.66 per cent carbon quenched in cold water from 1050° C. came out about two-thirds austenite and one-third martensite. Carbon steels of 1.2 per cent carbon and 0.83 per cent carbon, quenched in cold water from 800° C., were both

martensitic as quenched. Pieces of each of these steels were submitted in succession, after the quenching described, to the following series of tempering treatments: 100° C., six hours; 150° C., five hours; 200° C., four hours; 250° C., three hours; 300° C., two hours; 380° or 400° C., one hour; 450° C., one hour; 500° C., one hour; 600° C., one hour; 700° C., one hour. The 450° C. treatment was not always given. The quenched pieces were carefully examined and their properties measured. Then they were given the 100° C. tempering and again examined and measured. Each successive tempering

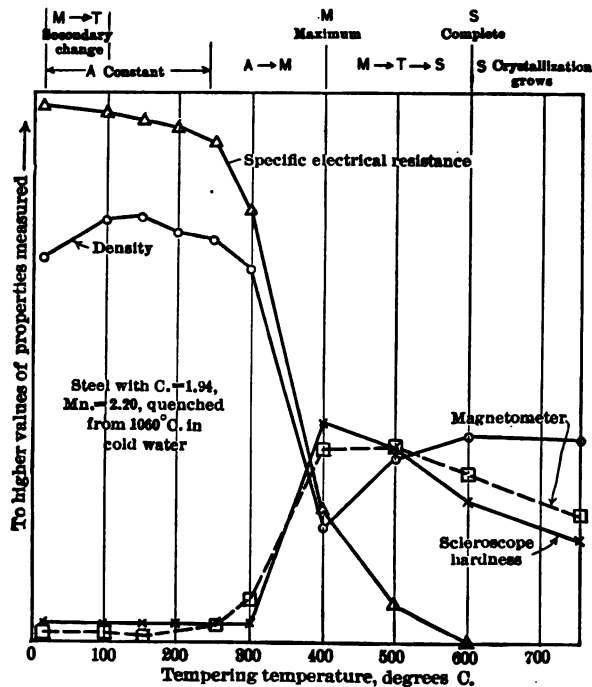


FIG. III.

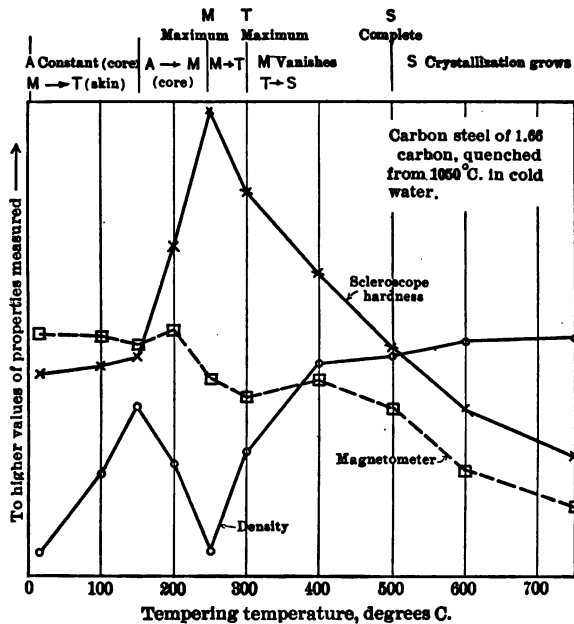
added its effects to all the preceding ones. The variation of properties is shown in Figs. III to III3.

Fig. III shows the changes in the steel which was entirely (?) austenitic as quenched. With carbon and manganese both very high this steel was very slow to change. Its microscopic appearance did not appreciably alter until it received the 300° C. treatment. The

400° C. treatment caused the bulk of the austenite to decompose to martensite. The 500° C. treatment changed a great part of the martensite to troostite and the 600° C. treatment left the steel in the sorbitic condition.

In the physical properties it will be noted that the austenitic structure is nearly non-magnetic, and is the softest structure of the series, while it has high density and electrical resistance. Minimum density and maximum hardness occur with the martensitic structure. The troostitic structure shows the greatest magnetic permeability.

The carbon steel of 1.66 per cent carbon (Fig. 112) changed in



quenching to the extent of forming about one-third of its structure as martensite. As the quenching conditions are practically the same as with the preceding piece, comparison of the two shows the effect of the carbon and manganese content in retarding the rates of reaction in heat treatment. The 100° and the 150° C. temperings cause in the 1.66 per cent carbon piece a puzzling increase in density, possibly explained by formation of troostite from martensite (a pressure

effect), while the austenite remains comparatively unaffected. The 200° and 250° C. temperings cause rapid decomposition of the austenite and formation of martensite. The hardness rises sharply and the density falls. The 300° C. tempering brings about the formation of maximum troostite, the hardness falling off and density rising. Sorbite becomes prominent in the 400° C. tempering, and in the 600° C. tempering the sorbitic structure is completely organized.

The 0.83 per cent carbon steel quenched from 800° C. (Fig. 113) fairly represents commercial practice with carbon tool steels. Inter-

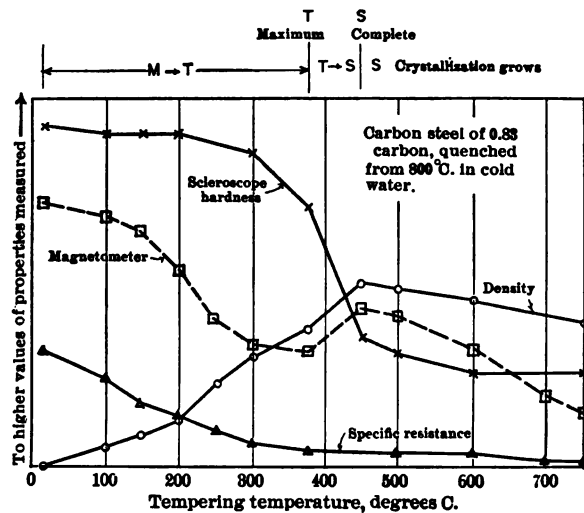


FIG. 113.

nal changes are so rapid that little austenite is retained by quenching. The significant facts of the tempering are the fall of hardness and the increase of density in the various temperings up to 450° C. This steel goes during quenching to some stage of reaction to the right of the maximum of the *B* (martensite) curve, Fig. 100; any tempering decreases the martensite. Sorbite formation was completed in Maurer's 450° C. tempering of this steel.

**293.** Of the changes in physical properties just discussed perhaps the most important in its effects on the practice of heat treatment is the change of density. Taking the density of the sorbitic structure as unity, troostite is about 0.998, pure martensite would be about



0.992, and austenite is about 1.008. The importance of the density changes will be seen if we compute the stress intensities required to cause similar changes of dimensions of the steel. A volume change of 0.001 part at ordinary temperature requires a pressure not far from 65,000 pounds per square inch. The volume change from austenite to pure martensite, if caused by pressure alone, corresponds to a pressure change of 1,040,000 pounds per square inch. Of course the steel must yield or break before reaching such a stress. It is obvious that transformation of austenite to martensite with the accompanying increase of volume is capable of setting up enormous internal stresses in the steel. This is the more apparent when we see by microscopic examination that the crystals do not all change simultaneously. If one crystal in a group changes from austenite to martensite, while the rest of the crystals remain as austenite, the crystal changing is put under heavy compression while it tries to thrust the surrounding crystals apart. Internal stresses set up in this way constitute one kind of **cooling stresses**. It is quite possible for the cooling stresses, set up during the quenching, to be as great as the strength of the metal. The metal would then have left no available strength to carry external load. **The available strength of a heat-treated piece is always the difference between the natural strength and the cooling stresses.** The piece may even be cracked or broken during quenching.

On account of the different densities of the allotropic forms of iron, heating and cooling stresses always occur as the steel passes through the critical temperature. Different parts of a piece, and the outside and inside of the cross-section of a piece, can never be quite at the same temperatures while the piece is heating or cooling. During the passage through the critical temperatures, therefore, internal stresses must always be set up in the material because of the temperature differences of various parts and differences of density or volume in the allotropic forms of iron. The change in allotropic form does not take place simultaneously in all parts of a piece.

In the quenching of a piece cooling stresses are unavoidably set up. The intensity of the cooling stresses is greater the more rapid the cooling given. It is the presence of these cooling stresses which makes it impossible to use hardened tool steel in the condition as quenched.

**294. Cooling stresses are relieved by tempering.** The relief of cooling stresses begins at temperatures as low as  $100^{\circ}\text{C}$ . and is rapid at  $200^{\circ}\text{C}$ . It is probable that cooling stresses are not completely removed from a quenched steel unless the tempering reaches a temperature of  $400^{\circ}$  to  $500^{\circ}\text{C}$ . Tempering accomplishes two things simultaneously. It relieves the cooling stresses and it allows the series of changes from austenite toward sorbite to make progress. Ordinarily the steel becomes softer during tempering because of disappearance of martensite. These two actions in tempering cannot be separated. Indeed the relief of cooling stresses is largely due to change of martensite to troostite, removing volume differences which are one cause of cooling stresses. Cooling stresses between different parts of a piece, due to differences of cooling of those parts, are relieved only by raising the temperature until the yield point becomes low enough to allow permanent sets.

**295.** The effectiveness of a quenching agent depends upon the rate of cooling it gives the metal throughout the "critical range" from the critical temperature down to about  $500^{\circ}\text{C}$ . The latter temperature lies just below the edge of the visible red heat. The most common quenching agent is cold water. The effectiveness depends somewhat

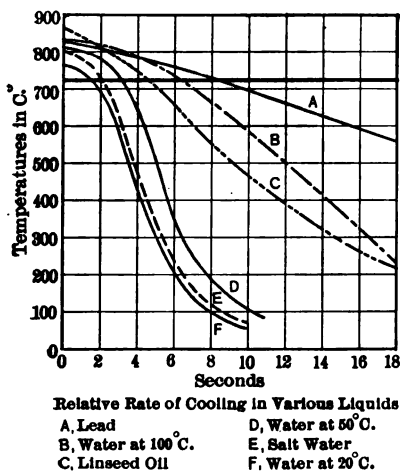


FIG. 114.

on the temperature of the water, being greater the colder the water. On this account brines have some advantage over water because they can be made colder without freezing. Water at its boiling point has practically no power for cooling red hot steel because a film of steam forms around the steel which keeps the water away from the steel. This steam formation causes even hot water to be very irregular in its quenching action if the temperature of the water exceeds  $50^{\circ}\text{C}$ . Any substance in solution in the water which will decrease the vapor pressure of the water at a given temperature will tend to stop the formation of the steam film. Such substances are  $\text{NaCl}$ ,  $\text{CaCl}_2$ , and many other

pressure of the water at a given temperature will tend to stop the formation of the steam film. Such substances are  $\text{NaCl}$ ,  $\text{CaCl}_2$ , and many other

salts. A cold brine is not much more effective than water of the same temperature, but the range of temperature over which brine is a dependable quenching agent is much wider.

Where less rapid cooling is desired during quenching various kinds of oils are used. In general the thinner or lighter the oil, the more rapid is the cooling it gives. Commercial quenching oils are generally very dark, heavy oils with high flash points. For cooling still less rapid than with the oils the hot steel may be plunged into liquid lead. Still slower cooling is obtained by exposing to a blast of air, or by allowing the piece to cool in still air.

Fig. 114, taken from data by Le Chatelier,\* shows the forms of actual cooling curves, temperature vs. time, of steel quenched in various ways. The pieces were rather smaller than those used in commercial work, making the cooling rates a bit too high to be representative of actual practice. The temperatures were taken with a thermocouple inserted to the center of the piece. The slopes of Le Chatelier's curves through the critical range of temperature give the following results in rate of cooling in degrees Centigrade per second.

Cooling Agent	Rate of Cooling (°C./per Sec.)
(Still air).....	(about 1)
Lead.....	17
Water at 100° C.....	45
Linseed oil.....	50
Mercury.....	65
Water at 50° C.....	115
Water at 20° C.....	140
Salt water.....	140
Water spray.....	145

The use of the water spray comes in commercially for pieces of large size or complicated shape, which could not be cooled simultaneously and equally in all their parts by plunging into a tank of water.

**296.** The nature of the quenching agent is only slightly more important than the effect of the size and the shape of the piece in determining the rate of cooling during quenching. The heat from the piece being quenched must escape through the surfaces of the metal,

\* *Revue de Métallurgie*, Sept., 1904.

while the amount of heat depends directly upon the mass of the piece. The rate of escape of heat from the piece, therefore, depends upon the ratio of surface to mass. In comparing pieces of the same shape it must be remembered that the surface varies with the square of the linear dimensions while the mass varies with the cube. Therefore, the ratio of surface to mass varies inversely as the linear dimensions of pieces of similar shape. **In general, the larger the piece, the slower is the cooling that can be obtained in any given quenching agent.**

297. In the table, given in Section 295, of cooling rates obtained with different quenching agents, it will be noticed that there are considerable steps in rate of action in changing from one agent to the next more effective one. The result of this in practice is that it is rarely possible to find a quenching agent which will bring the *A-M-T-S* transformation during quenching exactly to the point at which we wish to stop it. It is, therefore, generally necessary to use a quenching agent which is a little too effective, and then qualify the result by tempering. Since the *A-M-T-S* changes during tempering take place at much lower temperature than those during quenching, the rate of action is slower and consequently control of amount of action is far more precise. The action of quenching is precise in the sense that the results can be duplicated; but the amount of action cannot be graduated to suit our desires. The amount of action in tempering can be so graduated. In the use of a quenching agent which is too effective, with the intention of qualifying the result by tempering, one danger must be avoided. The more rapid the quenching, the more severe are the cooling stresses. While the tempering more or less removes the cooling stresses, it cannot heal up microscopic cracks once they have been formed. Too drastic quenching must be avoided. An attempt should be made to find the cooling agent which comes nearest to doing away with the necessity of tempering.

298. Attention must now be called to a peculiarity in the shape of the cooling curves in Fig. 114 and its consequences in heat treatment. It will be noticed that the pieces do not begin to cool at their maximum rate until some time after the plunging of the piece in the quenching agent. The reason for this action is in the method of escape of the heat from the metal. The outer skin of the metal may be considered as cooled almost instantaneously to the temperature of the boiling

point of the quenching agent if that be water or oils, or to the temperature of the quenching bath if that be molten salts or metals. The heat from the central portions of the piece is then carried out by metallic conduction to the outer skin. It takes an appreciable time to establish the temperature gradient of the heat flow. Until this gradient is established the cooling of the central parts of the piece is slow.

In order to make the quenching effective, the metal must have begun to cool at its maximum rate before its temperature goes below the critical; and the maximum rate of cooling should be maintained through the critical range of temperatures. **In the heating of the metal preparatory to quenching it is necessary to reach a temperature far enough above the critical temperature so that in quenching the maximum rate of cooling will have been attained before the critical temperature is passed.** Otherwise the quenching loses markedly in effectiveness because the cooling is slow just at the point where rapidity of cooling is needed.

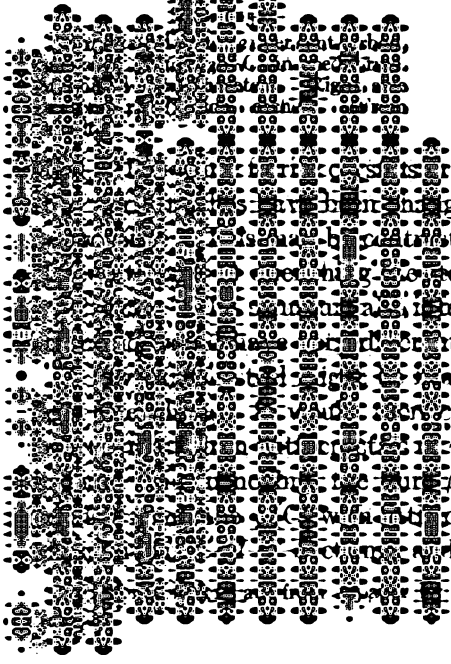
**299.** On the other hand, the heating before quenching must not go too far above the critical temperature because of the danger of crystal growth of the austenite. The higher the temperature above the critical, the greater is this danger. As a compromise between the need of effective cooling and the danger of overheating, the quenching temperature will generally be found in commercial work between  $50^{\circ}$  and  $150^{\circ}$  C. above the critical temperature of the particular steel. The smaller the piece, the less is the margin needed, because thinner metal requires less time to establish the temperature gradient at the beginning of cooling.

**300.** So far in this chapter we have assumed that as a preliminary of heat treatment, the steel had been heated to bring it into the austenite field of the equilibrium diagram, Fig. 94. It is not entirely true that no results can be obtained by rapid cooling if we do not start with the pure austenite. We must have at least some austenite present at the beginning of the quenching.

If we heat the steel into the field of  $\alpha + \gamma$  (C) it will consist of a mixture of ferrite and austenite crystals. The ferrite crystals will be nearly pure iron while the austenite crystals will contain about 0.8 per cent carbon. In quenching, the ferrite will be unaffected; the



# CONSTRUCTION



the  $A-M-T-S$  change. The temperature of the austenite  $\gamma$  (C) field. Therefore, the quenching is rather low and probably pass through the quenched piece will show a traces of troostite. Fig. 115\* steel from the  $\alpha + \gamma$  (C)



FIG. 116.

Carbon steel with 0.21 per cent carbon quenched from 1150° C. in iced brine. Magnified 460 diameters. (From Arnold.)

the same as before the treatment. the treatment from pearlite with Fig. 116, showing the very pearlite is a point entirely within the troostite and troostite but is very much

to the field of  $\beta + \gamma$  (C) because of a mixture of crystals of containing about 0.3 per cent carbon transforms almost instantaneously in solid solution goes rapidly to go to troostite even in

Arnold, in *Iron and Steel Inst. Jour.*,

is a mixture of ferrite and pearlite. In turn from the  $\alpha + \gamma$  (C) transformation, which could result from the latter transformation,  $\alpha + \text{Fe}_3\text{C}$ , we will be dealing with a mixture of more of carbon in solid solution. The quenching will affect the changes are concerned. The use of the high carbon content pearlite is very likely to be the result of quenching.

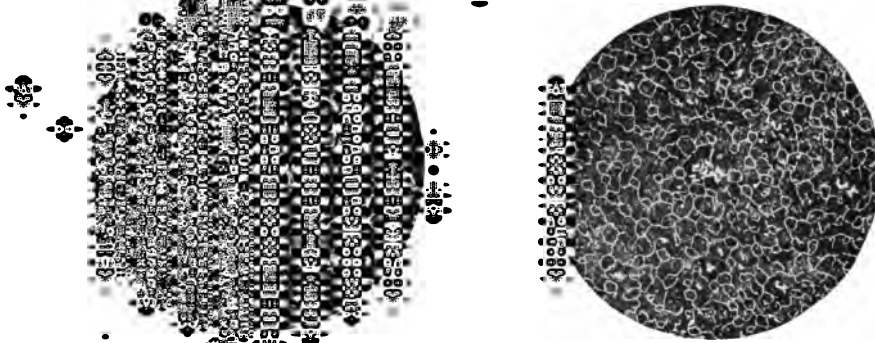


FIG. 118.

Carbon steel with 1.78 per cent carbon, quenched from 975°C. in water. Magnified 460 diameters. (from Arnold.)

The result is a martensite structure, which is that the cracking of the steel probably spoil this steel. In another possibility in heat treatment, the heating preliminary to the quenching where the solubility of carbon in austenite is high. This leads to a disappearance of the austenite and an increase in the carbon in solution. This increase in carbon content is the response to heat treatment,

## CONSTRUCTION

Fig. 118 shows the result of the treatment shown in Fig. 117 quenched to 1000° C. Cementite which had not been dissolved is still visible. The fine white lamellae of cementite which separated from the austenite during quenching, before the tempering had become rapid. The structure is shown in the right-hand half of Fig. 118.

This is a very puzzling phenomenon which is the presence of the carbon as graphite in a steel which comes from a cast steel to which unusual amounts of silicon were added to make sound. The carbon content is also abnormally high, about 1.8 per cent. In the quick cooling of the casting, as in the casting of white cast iron, the structure formed during freezing is  $\gamma$  (C, Si). This is shown in the right-hand half of Fig. 118. Although of course the  $\gamma$  (C, Si) had transformed to pearlite at the critical temperature, the zig-zag white crystals are of cementite which formed during freezing and comes down to 1000° C. This casting was reheated to 1000° C., held for about three days, and then cooled to 750° C. for about three more days. The structure obtained is shown in the left-hand half of Fig. 118. It analyzed 1.38 per cent graphite and 1.0 per cent combined carbon. The background is pearlite, containing more than 1.0 per cent carbon. Bedded in this are peculiar "knots" of ferrite and cores of graphite with rims of ferrite. To understand what has happened we must go to the equilibrium diagram, Fig. 79. The structure at 1000° C. is austenite plus graphite. This is unstable. Silicon displaces carbon from the  $\gamma$  which  $\gamma$  can hold in solution at 1000° C. According to the equilibrium diagram, but is reduced by the presence of silicon. Carbon not in solid solution becomes graphite. On cooling through the  $A-M-T-S$  changes, the steel deposits practically all of its carbon with the formation of graphite.

A similar action may happen in any steel of high carbon content above 1000° C. Cementite being unstable with graphite formation. The white carbon may be partially recombined with the steel between 750° and 800° C.



This is not effective because the cementite formation is slow. The other is to heat the steel up to about  $1100^{\circ}\text{C}$ ., when graphite will be absorbed into the austenite solid solution until the carbon content of the solid solution is 1.6 per cent, less the displacing action of the silicon. The only thoroughly satisfactory way to recombine the graphite is to remelt the steel.

**302.** Ordinarily we assume that the chemical analysis of the steel is the same throughout. Very commonly, however, this is not the case. Oftentimes in the heating of steel before quenching the furnace gases have an oxidizing effect upon the steel and more or less completely decarbonize the outer surface. In cases where the surface hardness of the steel is important we must avoid this decarbonization or must so design the piece as to allow for grinding off the decarbonized surface, which will naturally be soft.

**303.** In many cases we wish the piece, as a whole, to have high strength with considerable ductility, and yet to have a very hard outer surface so as to resist wear. An automobile transmission gear, or cam shaft, may serve as examples. For this work we take stock metal of low carbon content and "**case-harden**" the outer surfaces. This consists in raising the carbon content of the outer surface until it is like tool steel. Then a heat treatment which will strengthen and toughen the low carbon core will make the high carbon case very hard. This possibility of case-hardening depends upon the fact that  $\gamma$  iron under proper circumstances will absorb carbon from a surrounding gaseous and carbon containing atmosphere (or carbon containing liquid, such as KCN). This atmosphere is usually a mixture of CO and  $\text{CO}_2$ . If there is too much  $\text{CO}_2$  we get oxidation instead of carbonization. The gases can penetrate the surface of the hot steel to a depth of about  $\frac{1}{8}$  of an inch. This penetration increases somewhat with temperature. To the depth to which the gases penetrate the carbonization goes on fairly rapidly, until the carbon content of the outer layer approaches saturation of the solid solution of carbon in austenite. This saturation limit is shown as a function of temperature by the line *OPB*, or *OPB'*, in Fig. 79. A second action, much slower, is the diffusion of solution carbon through the austenite from any place of high carbon content toward places of low carbon content. A certain amount of this diffusion action must be obtained in successful case-hardening. If there is a sudden change of carbon content between the case and core the difference of volume change during

quenching will cause the case to peel from the core. By allowing sufficient time for diffusion there will be a region of gradual change from the high carbon of the case to the low carbon of the core. This gives a distribution of the cooling stresses during quenching so that case and core do not separate.

Because of the difference of carbon content of case and core the proper quenching temperatures of the two are generally different. The heat treatment which is best for refining and strengthening the core will probably not be good for hardening and strengthening the case and vice versa. There is also great need for refining the grain in the case, for during the formation of the case the high-carbon austenite has become coarse grained, having been held probably for some hours, at least, at temperatures  $900^{\circ}\text{C}$ . or higher.

The heat treatment of the case-hardened object will generally consist of: (1) air cooling after carbonizing; (2) reheating to  $900^{\circ}$  to  $950^{\circ}\text{C}$ ., hot enough to take the core into the austenite condition, air cooling again for refining the grain; (3) reheat to about  $800^{\circ}\text{C}$ . and quench for hardening the case and toughening the core. If the cyanide process has been used and the case is thin and not very high in carbon the heat treatment is generally simplified, because less refining of the grain is necessary. The cyanide case-hardening does not need such high temperatures, or as long times of heating as the gaseous processes.

**304.** The more complete theory of heat treatment, especially that of the relations of the various transformation forms to each other, requires somewhat of an excursion into physical chemistry. A knowledge of the relations to each other of stable and metastable states must first be had.

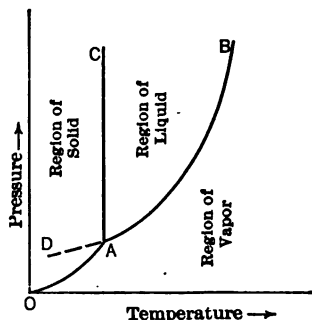


FIG. 120.

When a substance under given conditions of pressure, temperature, etc., has no tendency to change, the substance is chemically and physically **stable** under those conditions. When the external conditions are such that a substance ought to undergo a change, but despite its lack of stability the substance does not change, it is described as **metastable**. For instance, liquid water is metastable at temperatures below its freezing point at ordinary pressures. Fig. 120 shows the pressure and temperature relations of the substance, water, in the solid, liquid and vapor states. The curve AB is the familiar pressure temperature curve for steam in contact with liquid water. At temperatures to the right of AB liquid cannot exist; the water will be in the state of superheated steam. OA is the vapor pressure curve for ice. This is of the same general form as AB. AC is the curve

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showing the change with pressure of the temperature of the freezing point, where liquid turns to solid. If the liquid water be carefully cooled it can be carried to temperatures below the freezing point *A*. When water is thus "supercooled" its vapor pressure is found to lie on *AD*, which is an unstable extension of *BA*. The supercooled water is metastable. If ice crystals are dropped into it, or if it is stirred, it will crystallize immediately and suddenly. The latent heat evolved during the formation of the ice from the supercooled water will bring the temperature of the mass back to the freezing point *A*. Comparing the curve *AD* with the curve *AO*, we find that at a given temperature the vapor pressure of the metastable phase of a substance is higher than that of the stable phase.

305. The rise of temperature due to the evolution of latent heat in the crystallization of the supercooled liquid water is paralleled by a similar action in steels at the critical temperature. When steel in the  $\gamma$  (C) condition cools through the critical temperature, it ought to change to  $\alpha + \text{Fe}_3\text{C}$ . It does not always start this change on schedule. Instead it may supercool considerably, about  $10^\circ \text{C}$ . in low carbon steels to as much as  $50^\circ \text{C}$ . in high carbon steels. When the change does start there is a rapid evolution of latent heat which raises the temperature of the steel. The rise of temperature is so large that the red color of the hot steel becomes visibly brighter. This action of spontaneous reheating toward the critical temperature, after the supercooling, is called the "recalescence" of the steel. A similar action occurs in the heating of steel through the critical temperature.  $\alpha$  and  $\text{Fe}_3\text{C}$  persist for a time to higher temperatures than the critical. Then they suddenly change to  $\gamma$  (C) with an absorption of heat, cooling the steel. The "decalescence" is less marked than recalescence. This lagging of the changes in steels at the critical temperatures must always be remembered in commercial heat treatment.

306. Fig. 121 shows the pressure-temperature plot of the phase relationships of the substance sulphur. Sulphur differs from water in that it has, at ordinary temperatures, two allotropic solid forms. From their crystal habits these are known as rhombic and monoclinic sulphur. *OA* is the vapor pressure curve for rhombic sulphur, *AB* for monoclinic and *BC* for liquid. *AD* shows how the inversion temperature of the allotropic change from monoclinic to rhombic sulphur, or vice versa, changes with pressure. The critical temperatures in steels are subject to similar changes by pressure. Liquid sulphur may be supercooled just as water may. This supercooling gives the metastable curve *BGF*. Solid rhombic sulphur may be superheated. If this superheating is rapid enough the rhombic sulphur may be carried to the point *G* and there melted. *G* is an unstable melting point for rhombic sulphur. If rhombic sulphur is slowly heated it changes to monoclinic and the monoclinic melts at the point *B*, which is the point commonly known as the melting point of sulphur.

307. If from Fig. 121 we extract the vapor-pressure curves of the liquid and the two solid forms, we have the relations shown in Fig. 122. If the liquid is cooled slowly it will freeze to solid *I* at the temperature of *E*. On further slow cooling solid *I* will invert or change to solid *II* at the temperature of *B*. In slow heating the series of changes will take place in the reverse order. If the cooling of the liquid is more rapid, complications ensue. If the supercooling of the liquid below the temperature of *E* does not get as far as *D*, solid *I* forms in the freez-

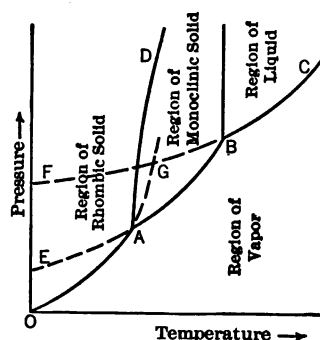


FIG. 121.

ing and subsequent events are about the same as in slow cooling. If the supercooling takes the liquid below the temperature *D*, then the solid *I* does not form during the freezing,

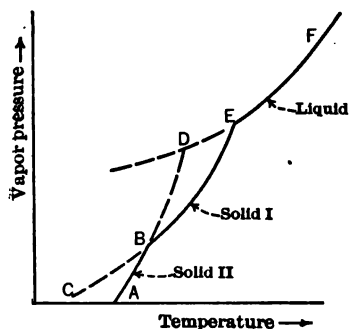


FIG. 122.

but instead the metastable solid *II*. In general the phase first formed during a change is the phase which at the temperature of the change has the vapor pressure next highest to that of the phase which is decomposing. In this case solid *II* formed from the liquid at temperatures just below *D* is metastable. If the temperature were kept between *B* and *D* solid *II* would change to solid *I*. The same circumstances which supercool the liquid would probably cause rapid cooling of the metastable solid *II*. If this solid *II* manages to survive without decomposition to below the temperature *B*, it will then have got into its region of stable existence.

**308.** It is quite possible for a substance to have a solid allotropic form which in the region of our experience is never stable. The necessary phase relationship is shown in Fig. 123. The vapor-pressure curves of the two solid forms do not cross as they do in Fig. 122. On slow cooling of the liquid phase of the substance in Fig. 123, freezing at the temperature *E* would give the stable solid *I*. This solid would come to ordinary temperatures without tendency to change. Supercooling the liquid to below the temperature of *D* would give the metastable solid *II*. With sufficiently rapid cooling this might be brought to ordinary temperatures. At ordinary temperatures it is possible that the rate of change from the metastable solid *II* to the stable solid *I* would be negligibly small for all practical purposes. An actual case of relations such as those of Fig. 123 occurs with minerals of the composition  $\text{MgSiO}_3$ . Four allotropic forms of this substance are known at ordinary temperatures. Only one form is stable. The other three are related to the stable form and to each other as solid *II* to solid *I* in Fig. 123.

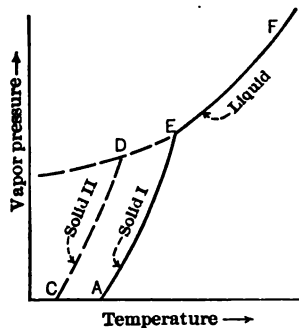


FIG. 123.

**309.** When two solid phases of a substance are related to each other as in Fig. 122, we may change either form to the other by carrying the temperature slowly enough through the transition point *B*. When solid *II*, stable at low temperatures, has been transformed into solid *I* by raising the temperature above *B*, the reverse change in cooling may be partially or completely suppressed by rapid cooling. This brings solid *I* to ordinary temperatures and corresponds to running down the metastable curve *BC* for solid *I*.

When the relations of two solid phases are as in Fig. 123, solid *II* can be obtained only by supercooling of the liquid. If solid *II* is held at high temperatures it will change irreversibly to solid *I*. Heating or cooling of solid *I* will cause no change except that a sufficient heat will melt solid *I*. There is no possibility of going from solid *I* to solid *II* without going through the liquid state.

**310.** An example of such action as that of Fig. 122 occurs in the cast irons. Molten iron-carbon alloys with more than 2.0 per cent carbon may freeze in two ways. If they are slowly cooled they freeze to  $\gamma$  (C) + G, austenite + graphite. This corresponds to

freezing at *E*, solid *I*, Fig. 122. In the equilibrium diagram, Fig. 79, this freezing temperature of the solid *I* is the temperature of the line *BH*. If the same liquid cast iron is quickly cooled it forms  $\gamma$  (C) +  $\text{Fe}_3\text{C}$ , austenite + cementite. This corresponds to freezing at *D*, solid *II*, Fig. 122. This freezing temperature shows as *B'F* in Fig. 79. The transition temperature between "gray" cast iron  $\{\gamma$  (C) + G, at freezing $\}$  and "white" cast iron  $\{\gamma$  (C) +  $\text{Fe}_3\text{C}$ , at freezing $\}$  is the line *PR* in Fig. 79, corresponding to the point *B* of Fig. 122. In actual practice the graphite formed in the stable (slow) freezing of cast iron does not change to  $\text{Fe}_3\text{C}$  at the transition temperature, but persists indefinitely in the metal. This corresponds to having cast iron cool, so far as the graphite is concerned, on the curve *EBC*, Fig. 122. White cast iron held at constant temperatures above *PR*, Fig. 79, changes irreversibly to graphitic cast iron. After this change has occurred the iron cannot be made again entirely white without remelting.

**311.** Stable and metastable phases differ in solubilities, that is, in the amounts required to saturate a solvent. At a given temperature a metastable phase of a system is more soluble than a corresponding stable phase. This relation must necessarily be true because the solution must be a possible path of change from the metastable to the stable condition at constant temperature. If quantities of both the stable and metastable phases are simultaneously in contact with a common solution, the solvent will absorb the metastable phase and deposit the stable.

An illustration of these differences in solubility is in Fig. 79 in the lines *PB* and *PB'* and also the lines *DE* and *D'E'*. *PB* is the solubility of the stable phase, graphite, in  $\gamma$  (C). *PB'* is the solubility of the metastable phase,  $\text{Fe}_3\text{C}$ , in  $\gamma$  (C). The solid solution  $\gamma$  (C) requires more carbon to saturate it when in contact with  $\text{Fe}_3\text{C}$  than when in contact with G, at the temperature where G is the stable phase. Below the temperature of *PR* the stability of  $\text{Fe}_3\text{C}$  and G is interchanged. Also in Fig. 79 *DE* is the solubility of the stable phase G in liquid iron; while *D'E'* is the solubility of the metastable  $\text{Fe}_3\text{C}$  in liquid iron.

**312.** The relations to each other of the allotropic forms of iron,  $\alpha$ ,  $\beta$  and  $\gamma$ , may be represented by a figure very similar to Fig. 122. The relation is shown in Fig. 124.  $\alpha$  is the

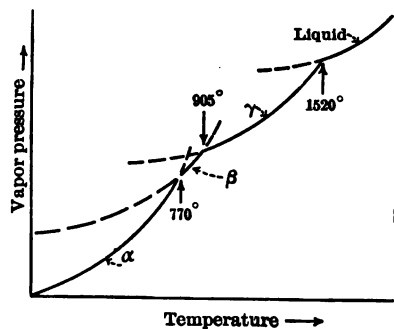


FIG. 124.

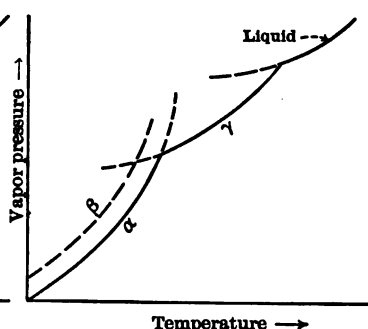


FIG. 125.

stable form from ordinary temperatures to 770° C.;  $\beta$  is the stable form from 770° to 905° C.; and  $\gamma$  from 905° C. to the melting point of the iron. The range of stability of  $\beta$  is very slight even in pure iron. The relation indicated by Fig. 125 is theoretically entirely possible. If Fig. 125 were true  $\beta$  would never be stable, but it would be obtained in heat treatment of steel, in a metastable condition, by quenching of  $\gamma$ . This suggestion is

of great theoretical importance, for it is a part of the explanation of the apparent mystery of the hardening of steel. If Fig. 125 were true,  $\beta$  would not appear at all on the equilibrium diagram, Fig. 79, except as do the metastable forms indicated by brackets [ ].

**313.** When carbon is taken in solution in the  $\gamma$  (C), the vapor pressure of iron from the solution is at first less than from pure  $\gamma$  of the same temperature. To represent the relations when carbon is present with the iron would require a three-dimensional plot with coördinates pressure, temperature and composition. By passing through this three-dimensional figure planes of constant composition, that is, parallel to both temperature and pressure axes, a series of figures will be cut out similar in general to Figs. 124 and 125. This series of figures may be superposed on one another in a single plane, yielding Fig. 126. It must be remembered in using Fig. 126 that the various  $\gamma$  (C) lines do not properly lie in the same plane. Since  $\alpha$  and  $\beta$  do not take carbon into solution, the  $\alpha$  and  $\beta$  lines in Fig. 126 are the same for all carbon contents. Change of carbon content affects only the  $\gamma$  (C) lines. To show the interrelation of Fig. 126 with the equilibrium diagrams, Figs. 79 and 94, the points *K*, *N* and *O* are used with the same meaning in all three figures. Fig. 126 shows, as do the equilibrium diagrams, that when the carbon content exceeds 0.6 per cent,  $\beta$  has no region of stable existence. Fig. 126 makes clear what the equilibrium diagrams do not show, that by supercooling to a sufficient degree  $\gamma$  (C) of any carbon content,  $\beta$  might be obtained at ordinary temperatures in the metastable condition, as a decomposition product of  $\gamma$  (C).

**314.** The vapor pressure of iron from  $\gamma$  (C) is to be expected to be a minimum for the composition of the solid eutectic, where carbon equals 0.9 per cent. Beyond 0.9 per cent

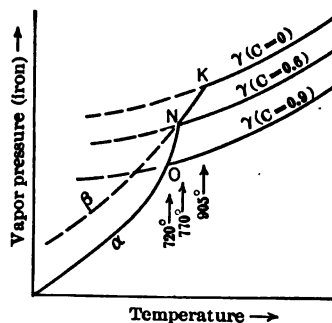


FIG. 126.

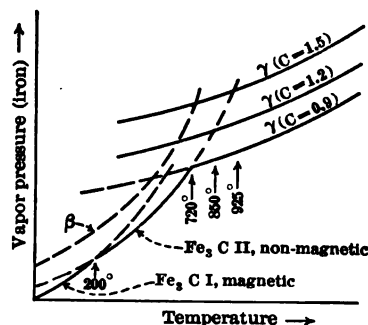


FIG. 127.

carbon the first phase regularly separating from  $\gamma$  (C) is  $\text{Fe}_3\text{C}$ . Fig. 127 shows in an incomplete form the phase relationships of the very high carbon steels.  $\beta$  has a metastable position with regard to all of these. Attention might be called to the two allotropic forms of  $\text{Fe}_3\text{C}$ .

**315.** Figs. 126 and 127 show that it is only in the case of the low carbon steels that the relation of  $\beta$  iron to  $\gamma$  and  $\alpha$  iron may be represented as in Fig. 124 with a region of stable existence. In the medium and high carbon steels the relation of  $\beta$  iron to  $\gamma$  and  $\alpha$  iron is always the metastable relationship shown in Fig. 125. When this metastable relation of Fig. 125 is true the change from  $\alpha$  to  $\gamma$  in heating will not be the reverse of the  $A-M-T-S$  change from  $\gamma$  to  $\alpha$  in the cooling used in heat treatment. Supercooled  $\gamma$  will yield  $\beta$  which later may change to  $\alpha$ , but in heating  $\alpha$  will change directly to  $\gamma$  without the inter-

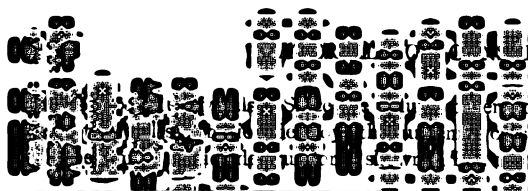
mediate formation of  $\beta$ . An important case of this irreversible action of the  $\alpha \rightleftharpoons \gamma$  change is to be found in the nickel and manganese alloy steels, and is brought into evidence by their magnetic behavior.

The difference in the results of heat treatment between low carbon and high carbon steels is also closely related to the differences in relations of  $\beta$  iron to  $\alpha$  and  $\gamma$  iron shown in Figs. 124 and 125. Fig. 124 represents the low carbon and Fig. 125 the medium and high carbon steels. If we supercool the  $\gamma$  iron with the relations in Fig. 125, we will obtain  $\alpha$  iron if the supercooling is slight, and  $\beta$  if the supercooling is greater. Sufficiently rapid cooling will always yield  $\beta$ , suppressing the formation of  $\alpha$ . Slow cooling will give  $\alpha$ , without intermediate formation of  $\beta$ . The difference between pearlite and sorbite is that pearlite forms by the direct change of  $\gamma$  iron to  $\alpha$  iron in slow cooling, while sorbite forms in more rapid cooling from the change  $\gamma$  iron to  $\beta$  iron to  $\alpha$  iron, with the consequent refining of structure. The alternating plate crystals of  $\alpha$  and  $\text{Fe}_3\text{C}$  in pearlite probably owe their form as much to their formation on cleavage planes of decomposing austenite as to the natural crystal habit of  $\text{Fe}_3\text{C}$ . Sorbite cannot be changed to pearlite without going back through the austenite field.

If we supercool  $\gamma$  iron with the relations shown in Fig. 124 (low carbon steel) slight supercooling will yield crystalline  $\beta$  iron which will almost immediately afterward change to  $\alpha$  iron because the cooling is slow. More rapid cooling, with more supercooling of the  $\gamma$  iron, will take the  $\gamma$  iron past the metastable extension of the  $\alpha$  curve, giving direct formation of  $\alpha$ , which will persist as the cooling continues. Theoretically a still more rapid cooling would again give us, from the supercooled  $\gamma$  iron, the formation of  $\beta$  iron, now non-crystalline, and the rapid cooling might retain the martensitic structure. Unfortunately, this cannot be realized in the low carbon steels. They either change from  $\gamma$  iron through crystalline  $\beta$  iron to crystalline  $\alpha$  iron, or from supercooled  $\gamma$  iron to non-crystalline  $\alpha$  iron to crystalline  $\alpha$  iron. We cannot cool the low carbon steels rapidly enough to get any other results and therefore we cannot harden the low carbon steels except for troostitic hardness.

**316.** In attempts to formulate explanations of the hardening of steel, controversy has generally raged around the part played by  $\beta$  iron. Often the very existence of  $\beta$  iron as an allotropic form of iron has been denied. The reasons for this are several. Crystalline  $\beta$  iron turns out to be very similar in crystal habit to  $\alpha$  iron, while  $\gamma$  iron is markedly different from the other two. That the density of pure crystalline  $\beta$  iron is less than that of pure crystalline  $\alpha$  iron at the same temperature has now been demonstrated. If curves are plotted against temperature for such properties of pure iron as magnetism, specific heat or electrical resistance, such curves are uniformly three branched, indicating that  $\beta$  has just as much right to recognition as an allotropic form as have  $\alpha$  or  $\gamma$  iron. The main trouble has been a lack of knowledge of the relations of stable and metastable forms such as have been illustrated in Figs. 120 to 127, and their discussion. A contributing factor has been the failure to recognize the  $A-M-T-S$  transformation as a sequence of overlapping chemical changes which may be represented as in Fig. 100. Another source of confusion has been an insufficient allowance for the difference in physical and engineering properties between crystalline and non-crystalline conditions of the various allotropic forms, or molecular kinds.

**317.** The starting point of the heat treatment of steel is the crystalline austenite structure which appears under the microscope as in Fig. 128. The austenite differs from the other crystalline forms of iron in the occasional presence of twinning in the crystals, which is a help in identification. The formation of martensite from the austenite may be considered as primarily due to the transformation of some or all of the  $\gamma$  molecules *in situ*



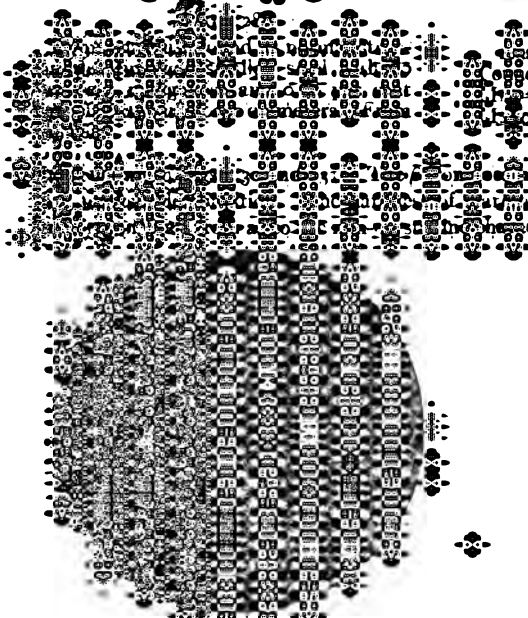
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the difference between  $\gamma$  and  $\beta$  is nearly one per cent. The transformation is more or less complete, and the final product, pure martensite.



FIG. 129.

martensite changing to martensite. The alloy is a manganese alloy steel with 1.45 per cent Mn, C = 0.120 per cent. Magnified 200 diameters. (From Guillet.)

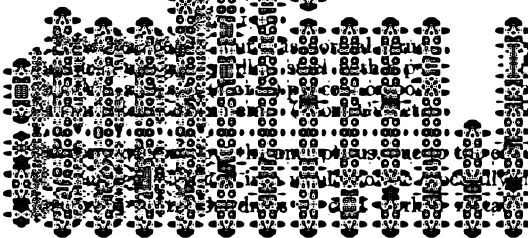


It does not crystallize when it is thus formed. This is not due so much to a special character of the structure, and its extreme



FIG. 131.

Pure martensite in steel properly quenched. Has been retouched with penciled lines to bring out boundaries of previous austenite crystals. Magnified 200 diameters. (From Guillet.)



due to the effect of the volume differences of  $\gamma$  and  $\beta$ . The hardness of martensite is great, but the amount of non-crystalline



the cold working, while its character, a deal with in the cold working of

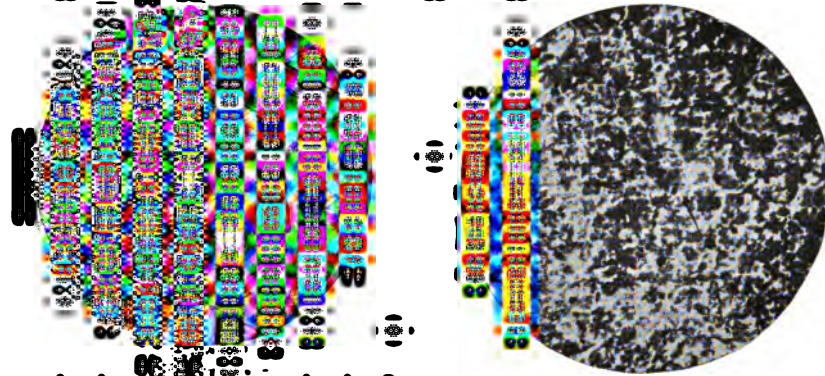


FIG. 133.

found as normal structure in a alloy steel with Mn = 5 per cent, per cent. Magnified 200 diameters. (From Wüst.)

different way the relation of the elements are taken into solid solution in

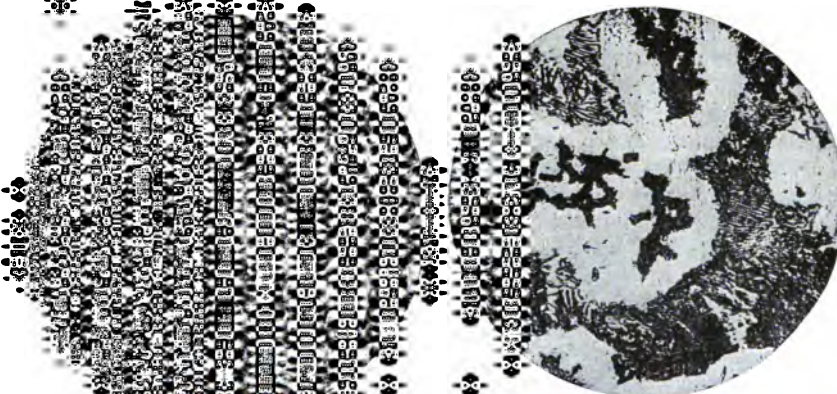


FIG. 135.

per graphite knot in malleable cast iron. Graphite surrounded by ferrite, in surrounded by pearlite. Magnified 200 diameters. (From Wüst.)

solid solution can occur when crystal elements sufficiently similar to build up together

in the same crystal structure. Whether the crystal units which bring carbon into solid solution in austenite consist entirely of carbon atoms, or of regular or irregular complexes of iron and carbon atoms, possibly chemical compounds, we do not know. The condition of the carbon in martensite is indicated by chemical analysis to be mainly that of solid solution. It may be considered that those  $\gamma$  molecules which have changed over to  $\beta$  have dropped their relationship to carbon and that the carbon in solution in martensite is held in a concentrated solution by the remaining  $\gamma$ ; while some portion of the carbon dropped from solution is in the free state in a condition of molecular subdivision.

**319.** The formation of troostite from the martensite is essentially a change of the non-crystalline  $\beta$  to non-crystalline  $\alpha$ . With this change the steel suddenly becomes magnetically soft, while the electrical resistance drops from the high values of austenite and martensite nearly to that of soft steel. Simultaneously with the transformation of  $\beta$  to  $\alpha$  the last of the  $\gamma$  in the martensite is disappearing. Neither  $\beta$  nor  $\alpha$  will hold carbon in solid solution. The carbon of troostite is found neither as solution carbon nor combined carbon. Some idea as to its condition may be obtained from the microscopic appearance of the steel, by working backwards from the known structures in cast irons, with high carbon content, to the steels with, comparatively, little carbon. Fig. 133 shows the appearance of troostite in steel. It is characterized by peculiar round knots, which are brought out rather better in Fig. 134. The cast-iron structure cited for analogy is the temper graphite knot, from malleable cast iron, shown in Fig. 135. In the making of "black-heart" malleable cast iron a large proportion of the temper graphite may appear at the same time with the appearance of magnetism, and the disappearance of austenite, just below the critical temperature. These circumstances are just those of the appearance of troostite in steels. Fig. 119 shows the appearance in a very high carbon steel of a temper graphite knot; the circumstances of its formation closely parallel the malleabilizing process in changing white cast iron to malleable. Figs. 134 and 133 show the same type of structure as Figs. 135 and 119, with differences such as might be expected from the much lower carbon content. Perhaps we might reverse the statement, and say that the formation of temper graphite below the critical temperature in malleabilizing is an exaggerated case of setting free of carbon in troostite formation, made evident by the large quantities of carbon released.

The genesis of the temper graphite knots seems to go back to either of two sources. It may be, at high temperatures, 900° to 1100° C., due to the decomposition of  $\text{Fe}_3\text{C}$ . Even then the graphite core is, at that time, probably surrounded not by pure iron, but by austenite saturated with carbon in solid solution. The alternative action, which is also the final action after  $\text{Fe}_3\text{C}$  decomposition, occurs at the end of austenite decomposition below the critical temperature. As the  $\gamma$  molecules change to  $\beta$  the carbon in solution concentrates in the remaining  $\gamma$  in spots, probably the cores of crystals. Then the last of the  $\gamma$  changes over, yielding pure  $\beta$  iron and carbon set free in some form. As  $\beta$  iron will not dissolve carbon we obtain the carbon free iron rim around the deposited carbon. Later the  $\beta$  changes to  $\alpha$ , completing the final form of the "knot."

It seems probable that we may assume the carbon condition in troostite to be that of temper graphite. In the troostite of steels the free carbon is small enough in amount, finely enough divided and well enough distributed, so that it is capable of recombination with the iron. The formation of free carbon at the end of austenite decomposition is another case of the well-known action, sometimes called a law, in physical chemistry, where a solution in decomposing deposits first the metastable form, which later converts to the stable. The equilibrium diagram, Fig. 79, shows that below the critical temperatures

graphite is the metastable and  $\text{Fe}_3\text{C}$  the stable product of the decomposition of the solid solution, austenite.

**320.** The formation of sorbite from troostite may be considered as consisting of two simultaneous actions. One is the chemical combination of the  $\alpha$  iron of the troostite with the adjacent carbon, forming the stable crystalline product  $\text{Fe}_3\text{C}$ . The non-crystalline  $\alpha$  iron, like "nascent" hydrogen, ought to be chemically very active. The other is a crystallization of the outstanding  $\alpha$  iron of the troostite. Both  $\alpha$  and  $\text{Fe}_3\text{C}$  crystals of the structure then grow in size if the temperatures and times of exposure permit. The sorbitic structure is chemically and physically stable at ordinary temperatures.

**321.** In summary, we may state the nature of the successive conditions of steel in heat treatment to be: (A) austenite, crystalline  $\gamma$  iron with carbon and other elements in solid solution; (M) martensite, a mixture of non-crystalline  $\beta$  iron with residual  $\gamma$ , the carbon and other elements remaining mostly in concentrated solution in the last of the  $\gamma$ ; (T) troostite, the iron as non-crystalline  $\alpha$ , the carbon free, and other chemical elements probably entering in solid solution in  $\alpha$  iron; (S) sorbite, crystalline  $\alpha$  iron, crystalline  $\text{Fe}_3\text{C}$ , with the other chemical elements mostly in solid solution in the  $\alpha$  iron.

Austenite is of fair strength, comparatively soft and ductile. Martensite is notable for its extreme hardness and high strength; it is usually brittle and its natural strength is often not available because of cooling stresses. Troostite is of excellent strength and fair ductility. As to hardness troostite is intermediate between martensite and sorbite, changing its relative position with the method of measuring hardness. By the scleroscope method troostite approaches martensite in hardness, while by the Brinell test it approaches sorbite. Cutting hardness is better measured by the Brinell method, for it is shown by experience that we need martensite instead of troostite for cutting hardness. The sorbitic steels are, on the whole, the lowest in strength of all the structures and they are not particularly more ductile than the austenites. At high carbon contents the ductilities of the austenites are higher than those of sorbitic steels. The relative strengths and

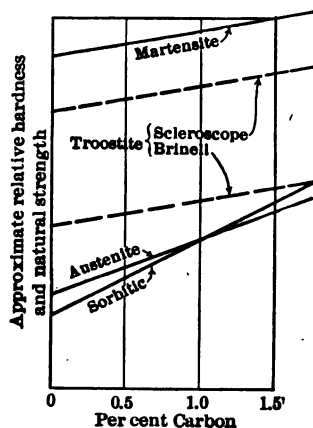


FIG. 136.

hardness relations are shown as functions of carbon content in Fig. 136. If this figure is to be used for relative yield point strengths use the scleroscope curve for troostite; if for relative maximum strengths use the Brinell curve. The figure is not intended to be very accurate and is more nearly true for hardness than for strengths.

It must always be remembered that in heat treatment only austenite and sorbite of the *A-M-T-S* series of structures can be obtained unmixed with preceding or following structures. So-called martensite and troostite are always mixed structures.

**322. It must be pointed out that carbon in steel is not essential in heat treatment.** Pure iron changes its allotropic forms so rapidly in cooling that we cannot control them. Heat treatment is made possible by any chemical element which, going into solid solution in the austenite, slows up the decomposition of the austenite to such an extent that we can catch the intermediate structures by sudden cooling. Carbon is the cheapest and most common and most effective of the elements which may perform this function. It will be seen in a following chapter on alloy steels that nickel, manganese, chromium or tungsten may have the same action as carbon in the carbon steels in making heat treatment possible. **We can have martensites or troostites without carbon.** The essential thing is to obtain the iron in a certain condition of molecular kind and organization, non-crystalline or crystalline.

## CHAPTER XVI

### ENGINEERING HEAT TREATMENTS OF CARBON STEELS AND PROPERTIES OBTAINED

**323.** The heat treatments commercially applied to carbon steels come under three main heads: (1) Steel is "**annealed**" to remove working stresses or cold hardening, to soften the metal for easier machining, or to remove cooling stresses and refine the grain. (2) High carbon steels are quenched and tempered to obtain cutting hardness for tools. (3) Steels of any carbon content may be heat treated to improve their strength, or their ability to take deformation, or both. After case-hardening the heat treatment has the double object of hardening the outer case, and strengthening and toughening the core.

The oldest of all heat treatments is that for cutting hardness. Annealing for softening the steel has also been long used. Heat treatment for high strength is very recent so far as general use is concerned; and it has not yet more than begun to be recognized commercially in its full importance.

**324.** It has been shown in Chapter XV, on the theory of heat treatments, that it is generally necessary to take the steel into the austenite region in order to secure any notable results from the heat treatment. By reference to the equilibrium diagram, Figs. 79 or 94, it will be seen that the temperatures for austenite lie above  $700^{\circ}\text{C}$ . Until recently, and to a large extent still in common practice, the temperature of the hot steel has been judged by the eye from the color of the metal. The scale of temperatures corresponding to colors is approximately as follows:

Temper- atures $^{\circ}\text{C}$ .	Color
525	Just visible red
650	Dull red
800	Cherry red
950	Orange
1100	Full yellow
1200	Light yellow
1400	White

Proper heat treatment temperatures vary from nearly orange with low carbon steels to cherry red with high carbon steels. The judgment of temperature by color is subject to large errors. The lighting conditions, whether night time or day time, or whether the shop is well or poorly lighted, greatly influence the appearances. The relation between color and temperature is also much affected by the thickness of the oxide scale on the hot steel. The thicker the scale the duller is the color for the same temperature. The error from this cause alone may be  $100^{\circ}\text{C}$ . The skill of the observer has much to do with the estimation of the temperature by eye. The greatest discrepancy is in the definition of "cherry"; some observers think of cherry as a color approaching the orange, others as approaching the dull red. Unhappily this difference of definition is very important, because cherry red is the most common temperature used in heat treatment. It cannot be denied that a skilled workman, using the same furnace and the same material day after day, comes to secure excellent results and to duplicate those results in a satisfactory way. The alteration of any of the conditions, however, involves a period of experiment and adjustment before the product is again satisfactory.

It has been found that to secure and duplicate the best results in heat treatment, the temperature of the furnaces ought to be held always within  $\pm 20^{\circ}\text{C}$ ., and often  $\pm 10^{\circ}\text{C}$ ., of a certain temperature which is peculiar to the steel in use and the result desired. Once this definite temperature has been found by experiment, it should always afterward be duplicated. Where heat-treated steel has to be produced in great quantity, and the quality must be uniform, the use of pyrometers in the furnace for exact measurement of temperature is indispensable. Even the most skilled workman cannot in his judgment of temperature approach the accuracy of a good pyrometer; far less can he approach its reliability under varying conditions. The instrument has no psychology.

The temperatures to which steel should theoretically be brought in the heat treatment furnaces lie above the line KNO of the equilibrium diagrams, Figs. 79 or 94. Considering the line KNO as a straight line the relation of critical temperature and carbon content is given by the equation

$$T_c = 900^{\circ}\text{C} - 200 \times \text{per cent of carbon}.$$

Manganese also lowers the critical temperature so that the equation for the carbon steels becomes in practice

$$T_c = 900^\circ \text{C.} - 200 \times \text{per cent carbon} - 50 \times \text{per cent manganese.}$$

This relation holds up to one per cent of carbon, and one or two per cent of manganese. With more than one per cent of carbon the critical temperature is the same as for one per cent of carbon. However, if we desire to cause the absorption of excess cementite, we must heat the steel to above the line *OPB* of Fig. 79.

The excess of temperature of the heat treatment furnace above the critical temperature of the steel varies from  $25^\circ$  to  $50^\circ \text{C.}$  for small pieces, to as much as  $100^\circ \text{C.}$  for large pieces. Higher temperatures bring risk of overheating the steel, especially if it is high carbon steel. Lower temperatures will result in inefficient quenching, for the reasons developed in section 298.

The time of heating in the furnace is generally to be kept as short as possible, while yet giving time for the complete formation of the austenitic structures from the preceding structures. Time beyond this is generally wasted because it can cause only growth of the austenite structure. In annealing for softness, however, some growth of the austenite is desirable. With large or complicated pieces heating of the steel must be slow enough to avoid serious expansion stresses, and also the similar stresses set up by the volume change caused by the transformation of  $\alpha + \text{Fe}_3\text{C}$  to  $\gamma (\text{C})$ . In some cases it may be necessary to do the heating in two different furnaces, using one to bring the temperature slowly up to near the critical, and the other to finish the heating.

**325.** Annealing for removing working stresses, or the cold hardening caused by pressing, bending, punching, etc., consists usually of heating to a dull red and cooling in the air. The amorphous material formed by cold working begins to recrystallize at temperatures as low as  $400^\circ \text{C.}$ , and is completely crystallized at  $650^\circ \text{C.}$  The lowering of the yield point by rise of temperature, as shown in Figs. 83 and 84, makes it possible for stresses between different parts of the piece, due to deformation of the cold metal, to disappear more or less completely if the steel is heated nearly to the critical temperature. In any reasonable time of annealing there is no danger of the crystal growth of the  $\alpha + \text{Fe}_3\text{C}$  structures. Ferrite which has been cold stressed to just

beyond its yield point shows a tendency to rapid growth at a dull red heat, but there is no need in this kind of annealing to give it time for this growth to become serious. If it has been cold stressed far beyond its yield point the crystals have been broken down by the stressing

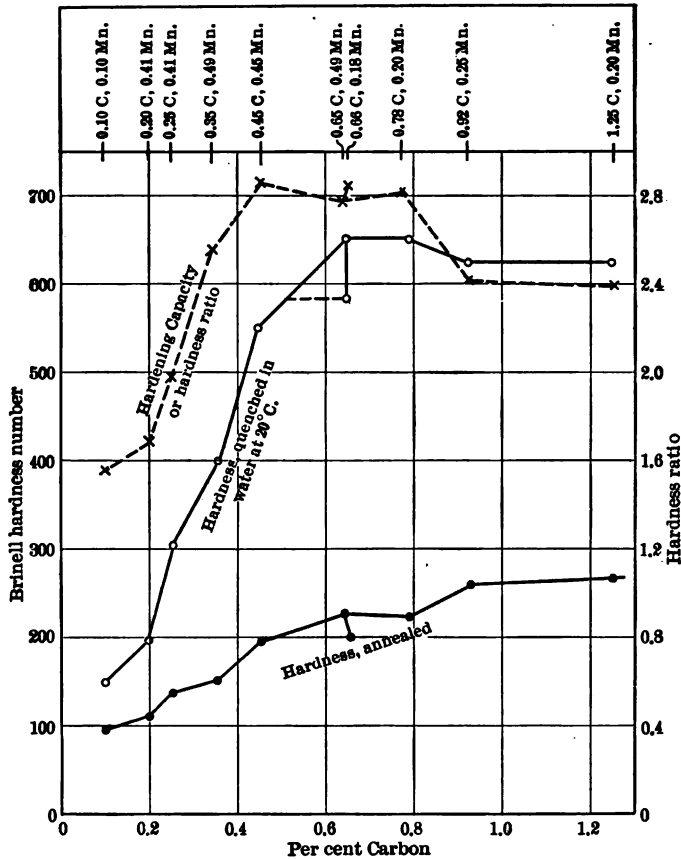


FIG. 137.

Hardening capacity of carbon steels. From data of Brinell and Wahlberg, *Iron and Steel Inst. Jour.*, 1901.

into smaller crystals and the regrowth at dull red will hardly take them back to more than their original size.

**326.** High carbon steels, in their normal condition, are often so hard and tough that they are difficult to machine. To them we apply



an annealing for softening which consists in heating them slowly well beyond the critical temperature so as to make a moderately coarse-grained austenite, then cooling very slowly in ashes, or lime, or in the furnace, so as to transform the coarse austenite into a coarse-grained pearlitic structure. (See Section 261.) This coarse-grained pearlitic steel is the softest we can secure for machining. It is, of course, understood that the piece is later to be heat treated to refine the grain.

**327.** Steel castings, and occasionally steel forgings, are coarse grained, and if complicated carry cooling stresses. To these we apply an annealing, which is essentially a heat treatment for refining the grain as explained in Section 261.

**328.** Heat treatment for hardness is restricted in the carbon steels to those of more than 0.5 per cent carbon. The reason for this will be apparent from Fig. 137, which shows the hardness obtained in water quenching as a function of carbon content. A cutting tool requires a hardness generally above 500 on the Brinell hardness scale, or 50 on the scleroscope scale. Tool steels rarely go above 1.25 per cent carbon because cooling stresses and other difficulties of heat treatment increase with carbon content. The reason why the hardness obtained in water quenching, and the hardening capacity, begin to fall off above 0.8 per cent carbon, in Fig. 137, is that we then begin to bring down some austenite with the martensite during the quenching. In such a case, however, the steel will get harder at first in tempering, or in use, while a steel which in quenching retains no austenite becomes softer by any tempering applied.

It has been mentioned that manganese is about half as effective as an equal amount of carbon in causing response of the steel to heat treatment. This is rather neatly shown in Fig. 137, where two steels of nearly the same carbon content, 0.65 and 0.66 per cent, differ by 0.31 per cent in manganese. It is evident that the hardening of the steel lower in manganese corresponds to that of the one higher in manganese if the carbon content of the latter were lowered by 0.15 per cent.

**329.** Fig. 138 shows the effect of quenching temperature upon hardness for different carbon contents. The quenching agent is cold water. All of these steels act alike in that too high a quenching

temperature involves crystal growth of the austenite in the heating before quenching. Coarse-grained austenite is a little slower to break down in heat treatment than fine-grained austenite; thus coarse grain increases the chance of austenite retention in quenching. A coarse-grained structure is also weaker than a fine one, and a coarse-grained austenite makes a coarse-grained martensite. With the higher carbon steels in Fig. 138, the crystal

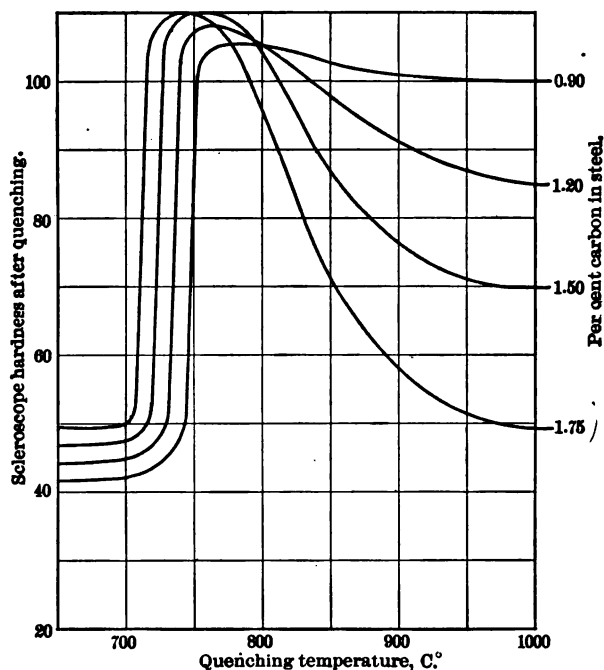


FIG. 138. Taken from Shore, *Scleroscope Book*.

growth of austenite at a given temperature increases with the carbon content. A factor of more importance is the increasing solubility of  $\text{Fe}_3\text{C}$  in  $\gamma$  (C) with rise of temperature, shown by the line *OPB* in Fig. 79. As the austenite increases in carbon content it breaks down more and more slowly during heat treatment, so that the 1.75 per cent carbon steel, for example, when quenched from  $1000^\circ\text{C}$ ., must retain about one-half of its structure of austenite unchanged, the other half going to martensite. The structure would look like that

shown in Fig. 118; the thin cell walls of cementite and the cooling stresses would probably make the steel hopelessly brittle.

It is evident from Fig. 138 that the range of quenching temperatures which gives the maximum hardness after quenching is very narrow for any one steel, and also changes with the composition of the steel.

**330.** While in commercial practice water is almost the only quenching agent used for hardening tool steel, it seems desirable to raise the question whether some slightly less effective cooling agent might not oftentimes be better. It seems probable that a light oil would, at least on small tools, give quite enough hardness, and with far less danger of injury of the steel by cooling stresses. The principles concerned are given in detail in Section 297.

**331.** Hardening of tool steels always overdoes its work, leaving the steel harder than necessary and also brittle and possibly weak from cooling stresses. Tempering is therefore necessary to make the steel again strong and tough. The amount of tempering action is generally judged by the "temper color." A portion of the surface of the hardened tool is cleaned of scale and polished. As the heating of tempering goes on a thin film of iron oxide grows upon this cleaned surface. The color of the oxide film is an approximate indication of the tempering temperature, or we might better say of the amount of tempering. (See Sections 288 to 290.) The relation usually given between temper color and temperature follows.

Temperature °C.	Temper color
225	Light straw
240	Dark straw
255	Yellowish brown
265	Reddish brown
275	Purple
285	Violet
305	Blue
330	Gray

Above 330° C. temper colors vanish. The metal then does not change in appearance until the dull red is reached. This table of temper color and temperature assumes a moderately rapid heating of the steel. If the metal is held at constant temperature, it will be found

that the temper colors change with time as shown in Fig. 139. In one way this makes the temper color rather more reliable than actual

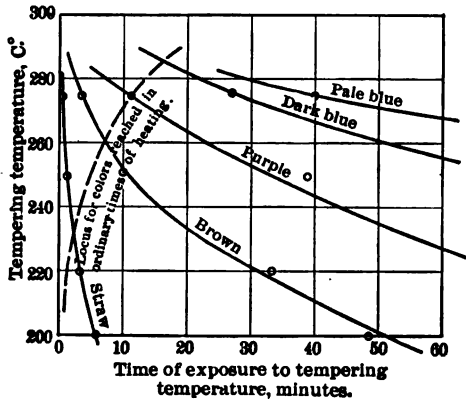


FIG. 139.

Relation between tempering temperature, time, and temper color. Data of Brearley.

temperature for commercial purposes, unless the time of the heating is also standardized in the commercial work.

Fig. 109 and its discussion show that the tempering effect varies with time at a given temperature. If it were true that the variation of temper color with time, at one temperature, paralleled the variation of the tempering effect with time, at the same temperature, then the temper color would be an

exact indication of tempering effect. The fact that the temper color does include a time effect as well as a temperature effect makes it an unusually good approximation in the commercial control of the amount of tempering. It is still true, however, in tempering as in quenching, that for the best results and continued duplication of results, temperatures must be accurately controlled by pyrometers and the time of exposure of the steel to those temperatures must be standardized. Fig. 109 shows that the lower the tempering temperature the more necessary is the control of time of tempering and vice versa.

**332.** It has been pointed out in Sections 286 and 290 that during quenching and tempering severe shrinkage stresses are set up between surface and core of the steel, at times having the outer parts in tension and the inner parts in compression and at other times reversing this stressing. These intense internal stresses search out very thoroughly any defects within the steel, the tension stressing resulting in cracking the piece on the defect. As examples of such defects we may cite piping or segregation of the ingot, either of which will lead to cracking in the center of the piece; blowholes in the ingot, which roll out into cracks or lines of weakness near the surface, which will become definite cracks by the action of the shrinkage stresses; rolling cracks due to

working too heavily and too cold, which will be opened by the shrinkage stresses.

Another cause of failure, which brings very large shrinkage stresses into play, is the choice of too low a carbon content for a moderately large piece of material. Small pieces cool fairly uniformly throughout their thickness, so that they are uniformly hardened from surface to center. Large pieces are cooled more effectively in quenching at the surface than at the core. If the carbon content is too low this difference in rate of cooling between surface and center may result in hardening the surface without hardening the core. Due to the volume difference between martensite and troostite or sorbite, the shrinkage stresses and cooling stresses are in this case unusually severe.

Considerable differences of rate of cooling in quenching or of rate of heating in tempering occur between surfaces, protruding corners, and reëntrant angles in the same piece. The protruding corner is cooled considerably more rapidly, and a reëntrant angle considerably less rapidly, than a surface. The corresponding differences in shrinkage, and possibly also in amount of martensite formation, cause a strong tendency to crack off the projecting corners and to start cracks in the material inward from the reëntrant angles. It is obvious that to a large extent these troubles can be avoided by proper design of the piece, or if the design cannot be changed, objects of complicated shape may be quenched by a specially arranged spray of quenching liquid rather than by quenching in a tank.

**333.** The most recent, historically, and the least developed, commercially, of heat treatments of steels is that for high strength while still retaining enough of ductility or toughness or shock resistance to qualify the material for structural use or as a part of machinery. If we attempt no cold working of the steel in assembly, a ductility or reduction of area, in the tension test of the material, of 10 to 15 per cent is ample for any ordinary purpose. The shock resistance of a brittle steel with a breaking strength of 100,000 pounds per square inch is  $\frac{1}{2} \cdot \frac{P_B^2}{E} = \frac{1}{2} \cdot \frac{(100,000)^2}{30,000,000} = 166.7$  inch pounds per cubic inch (resilience). A troostitic steel with a yield point of 100,000 pounds per square inch and ductility of 10 per cent has a shock resistance of approximately  $100,000 \times \frac{1}{10} = 10,000$  inch pounds per cubic inch.

The enormous increase in shock resistance by raising the ductility from nothing to 10 per cent shows the advantage and the necessity of moderate ductility in steel for structural or machine parts. Further increase of ductility would, with the same yield point, increase the shock resistance still more, but is unnecessary. Most normal steels have more than 10 per cent ductility, but have yield points ranging from 40,000 to 80,000 pounds per square inch. The working stresses cannot exceed one-third to one-half the yield point stress. If we use a steel of low yield point we must therefore put into a structure a large weight of metal in order to obtain a given strength. The saving of weight and space by the use of material of high yield point may more than repay the higher cost of such material, and in some cases, the automobile, for example, it is necessary even regardless of the initial cost.

It is commonly assumed that any attempt to raise the strength of steel by heat treatment is compensated by a corresponding loss of ductility or reduction of area, so that the value of the metal is not greatly increased. Such action often occurs, but is by no means necessary; and the heat-treated metal is improved for actual use provided the ductility remains above 10 per cent. It is quite possible to raise the strength by heat treatment without loss of ductility, or at least without making the steel approach brittleness. It is also possible to raise the ductility without loss of strength, especially in high carbon steels. The possibilities of various treatments in the carbon steels will be shown in the following discussions of Figs. 140 to 144.

In these figures the first part of each figure is from data in the paper of Brinell and Wahlberg, *Jour. of the Iron and Steel Inst.*, 1901. The second part of the figure is from data on a similar steel, taken in laboratory class work in Sibley College in 1914-1915 by E. H. Dix, Jr. His data is from averages of at least six pieces at each treatment, giving similar results, and showing that the results can be reproduced; Brinell and Wahlberg's data is from one piece for each steel and treatment. Brinell and Wahlberg applied the same arbitrary series of heat treatments to a considerable number of different steels. It was only by accident that they brought out the best results obtainable from any particular steel. Dix's work was directed to showing the best results obtainable from each steel, knowing the structures desired and the treatment needed to bring out those structures.

Fig. 140 shows the results on a structural steel. Annealing at  $350^{\circ}\text{C}$ . can have only the effect of removing working stresses and cold hardening. If the metal has been properly rolled the annealing will show no effect. Annealing at  $850^{\circ}\text{C}$ . means a partial passage above

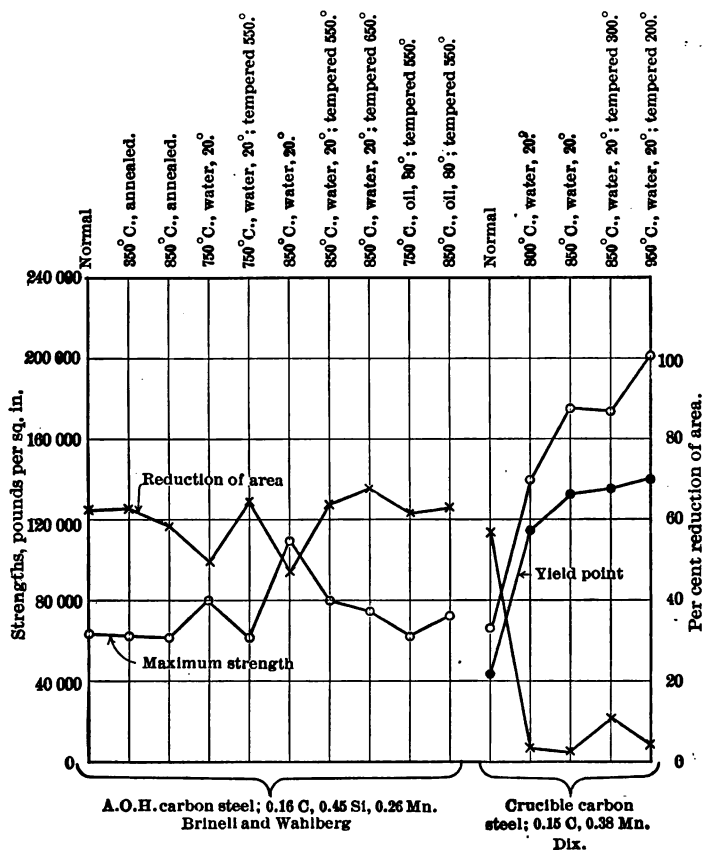


FIG. 140.

the critical temperature with some softening of the steel. Water quenching from  $750^{\circ}\text{C}$ . slightly hardens this steel while lowering its reduction of area. At  $750^{\circ}\text{C}$ . this steel consists of about 85 per cent ferrite crystals and about 15 per cent austenite crystals, the latter containing about 0.8 per cent carbon in solid solution. Quenching does not effect the ferrite, but the intermixed austenite crystals go through

martensite toward troostite. The quenching temperature is too near the critical temperature of the austenite to give the maximum effects. Tempering this quenched piece to  $550^{\circ}\text{C}$ . practically restores the normal structure and properties. Water quenching from  $850^{\circ}\text{C}$ ., where the steel is in the condition  $\beta + \gamma$  ( $C = 0.25$ ), with about three-fourths of the crystals  $\gamma$  ( $C$ ), gives considerably more result than the  $750^{\circ}\text{C}$ . quenching, and the general refining of structure causes this result to persist somewhat in tempering even to  $650^{\circ}\text{C}$ ., though the structure goes back from troostitic to sorbitic. Oil quenching would not be very effective on this steel in any case, and tempering to  $550^{\circ}\text{C}$ . would undoubtedly remove any effects of such quenching.

In the equilibrium diagram, Fig. 94, it will be seen that the critical temperature of this steel would be about  $860^{\circ}\text{C}$ . in the absence of manganese; with its manganese it is about  $840^{\circ}\text{C}$ . Because the low carbon steels go through the  $A-M-T-S$  change very rapidly, water quenching is necessary to give an appreciable result in heat treatment. This gives the key to the treatments used by Mr. Dix. His steel should give more response to heat treatment than the other because the manganese content is higher and the critical temperature lower. The effect of raising the quenching temperature is very sharply evident in Fig. 140. It is also shown that the highest strength is not combined with sufficient reduction of area for good shock resistance unless the water quenching is followed by a slight tempering. The value of the heat treatment is shown by comparing the normal properties with those of the pieces quenched at  $850^{\circ}\text{C}$ . in water and tempered at  $300^{\circ}\text{C}$ . The normals have a yield point of 44,000 pounds per square inch with a reduction of area of 58 per cent. The treated pieces have the yield point at 135,000 pounds per square inch and a reduction of area of 11 per cent. As this deformation is ample for shock resistance, it is shown to be possible to raise the yield point of a structural steel to three times the normal without losing too much in ductility and shock resistance. The rise of maximum strength from 67,000 to 175,000 pounds per square inch is also nearly in a three to one ratio. It is possible that a  $950^{\circ}\text{C}$ . quenching in water with  $300^{\circ}\text{C}$ . tempering would have done still better. When the steel is quenched from a point entirely within the  $\gamma$  ( $C$ ) field, instead of from the fields  $\alpha + \gamma$  ( $C$ ) or  $\beta + \gamma$  ( $C$ ), the resulting internal structure is nearly homogeneous



instead of patchy, and the internal stresses (cooling stresses) are less serious.

Fig. 141 deals with a machine steel. Using the equation for finding critical temperature given in Section 324, it is found that the first steel

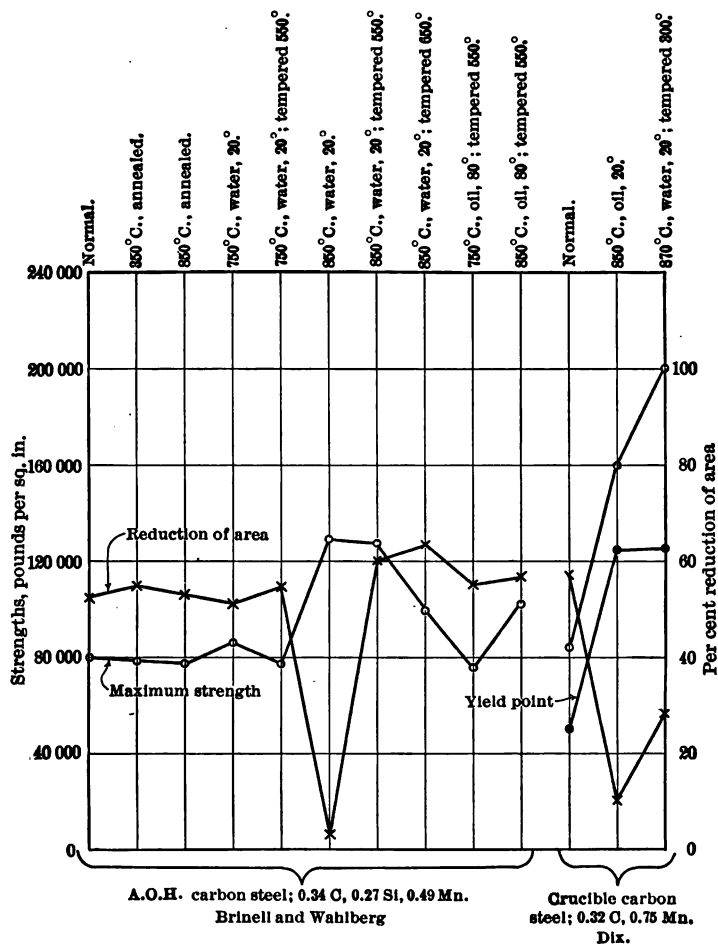


FIG. 141.

in Fig. 141 is entirely  $\gamma$  (C) above  $810^{\circ}\text{C}$ . and the second above  $800^{\circ}\text{C}$ . Quenching temperatures  $50^{\circ}\text{C}$ . higher than these or slightly more should give the best results for small pieces. This is borne out in Brinell and Wahlberg's results by the sharp effect of the  $850^{\circ}\text{C}$ .

quenching, and of that quenching combined with tempering. Tempering at  $550^{\circ}\text{C}$ . is too high, for it would undoubtedly pull the yield point well back toward the normal value. As with the structural steel, oil quenching of this machine steel will have its results practically nullified by tempering as high as  $550^{\circ}\text{C}$ . This is brought out by comparison of Brinell and Wahlberg's results with those of Dix. Oil quenching without tempering or water quenching with slight tempering will leave the machine steel with ample reduction of area while raising the strength to very high values. For comparison we tabulate Mr. Dix's results.

Properties	Condition of steel		
	Normal	$850^{\circ}\text{C}$ ., oil	$870^{\circ}\text{C}$ ., water, tempered $300^{\circ}\text{C}$ .
Yield point.....	53 000	124 000	124 000
Maximum strength.....	85 000	160 000	202 000
Reduction of area.....	56	11	29

It is probable that a slightly lower tempering temperature after water quenching would have left the yield point higher while still retaining sufficient reduction of area.

Fig. 142 shows the results obtained on a steel corresponding to railroad rail and tire steels. As before, the quenchings from  $750^{\circ}\text{C}$ ., with or without tempering, have little result. The quenchings from  $750^{\circ}\text{C}$ . are ineffective because the temperatures are too low to give an effective rate of cooling through the critical range of temperatures. The critical temperatures of these steels, above which they are entirely  $\gamma$  (C), are about  $760^{\circ}\text{C}$ . Quenching from  $850^{\circ}\text{C}$ . in water made Brinell and Wahlberg's test piece so brittle they could not test it. Another piece with the same quenching, but tempered to  $550^{\circ}\text{C}$ ., had its maximum strength at 169,000 pounds per square inch as compared with the normal 123,000, and reduction of area 33 per cent as compared with the normal 24. Increasing the tempering temperature to  $650^{\circ}\text{C}$ . brought the strength back down toward the normal but still further increased the reduction of area. The piece quenched in water from  $850^{\circ}\text{C}$ . was martensitic; tempering at  $550^{\circ}\text{C}$ . made it troostitic and at  $650^{\circ}\text{C}$ . made it sorbitic. The increase in deforma-

tion over the normal is probably due to greater internal homogeneity. Oil quenching from  $750^{\circ}\text{C.}$  would have little result in any case and the tempering at  $550^{\circ}\text{C.}$  would make even that disappear. The oil

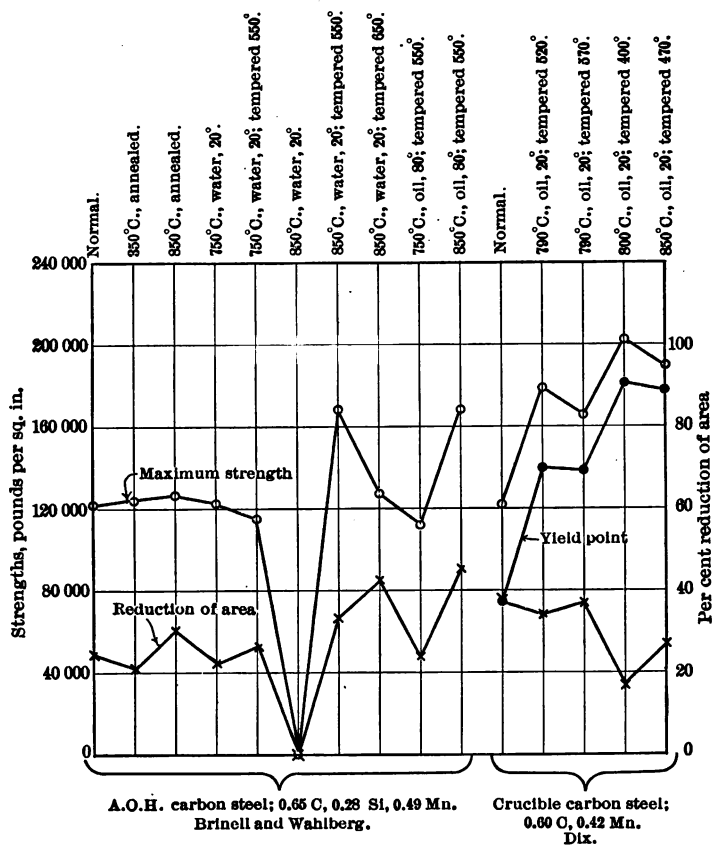


FIG. 142.

quenching from  $850^{\circ}\text{C.}$  should make this steel troostitic with perhaps a little martensite. Tempering to  $550^{\circ}\text{C.}$  would then produce a troostite-sorbite mixture, explaining the excellent results shown. Dix's work shows for a similar steel even higher strengths, while keeping the reduction of area more than high enough for structural purposes. From the yield point of 75,000 pounds per square inch, maximum strength of 123,000 and reduction of area of 38 per cent in the normal steel, the jump was made by quenching from  $800^{\circ}\text{C.}$  in

oil and tempering at  $400^{\circ}\text{C.}$ , to a yield point of 183,000, maximum strength 203,000 and reduction of area 17. The structure giving this is troostitic.

Fig. 143 shows Brinell and Wahlberg's results on a tool steel of 0.94

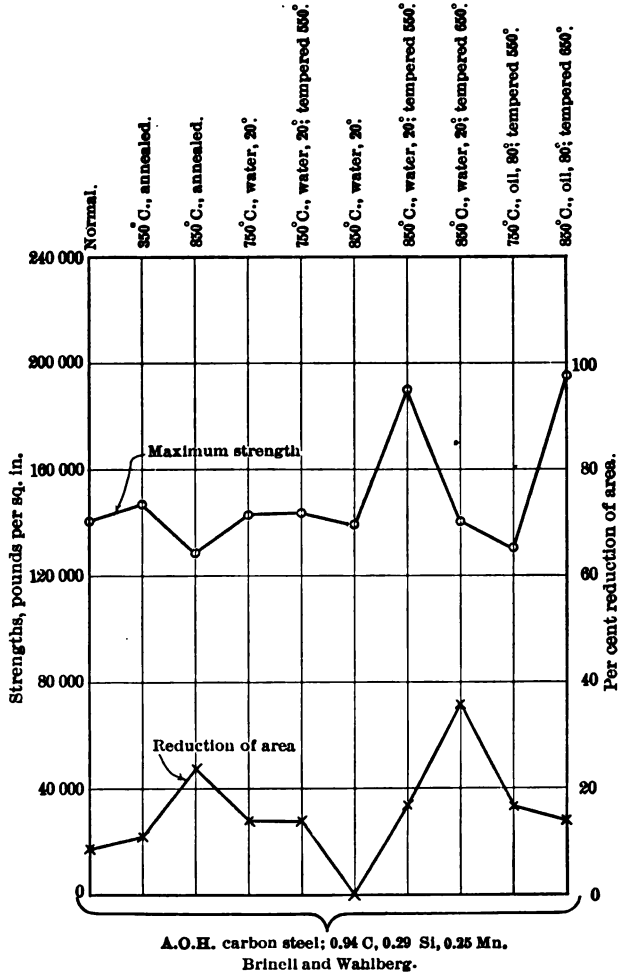


FIG. 143.

per cent carbon. The softening of the steel by the  $850^{\circ}\text{C.}$  annealing is now considerable. Again the  $750^{\circ}\text{C.}$  quenchings, with or without tempering, have little effect. The water quenching from  $850^{\circ}\text{C.}$

made the steel brittle; its strength was about normal, and would have been much higher but for cooling stresses. Tempering to  $550^{\circ}\text{C}$ . after this quenching gets rid of the cooling stresses and changes mar-

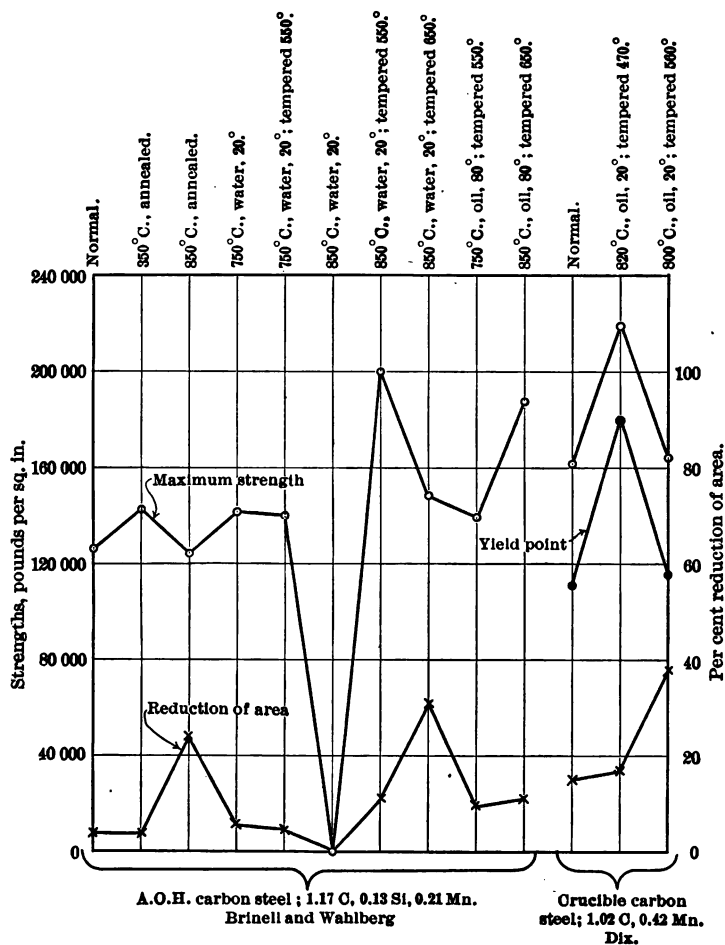


FIG. 144.

tensite into troostite, putting the strength up to 188,000 pounds per square inch, and the reduction of area, 14 per cent, is better than the normal, 8 per cent. Tempering to  $650^{\circ}\text{C}$ . makes the structure sorbitic, strength back to normal and reduction of area up to 35 per cent. The  $850^{\circ}\text{C}$ . oil quenching with tempering at  $550^{\circ}\text{C}$ . again gives a

very high strength and is undoubtedly a safer treatment than the water quenching.

Fig. 144 shows the results on steel of still higher carbon content. Dix's steel with 1.02 per cent carbon is really comparable with Brinell and Wahlberg's 1.17 per cent carbon because of the lower manganese in the latter case. Brinell and Wahlberg's results here entirely parallel those on the preceding steel. Dix's results are tabulated below, showing that it is possible in this steel to raise the strength markedly without loss in the reduction of area and vice versa.

Properties	Condition of steel		
	Normal	820° C., oil; temper 470° C.	800° C., oil; temper 560° C.
Yield point.....	110 000	183 000	114 000
Maximum strength.....	161 000	218 000	164 000
Reduction of area.....	15	16	38

In summary of the results which may be obtained in the heat treatment of carbon steels for high strengths, we will tabulate Dix's best results as to strengths where the reduction of area was kept above 10 per cent. The results quoted are not the best individual pieces, but are averages of six or more pieces giving similar results. After looking over these results the question will certainly occur whether it might not be better in commercial practice, before purchasing expensive alloy steels, to see what can be done by the proper heat treatment of the carbon steels.

Composition	Condition or treatment	Properties		
		Yield point	Maximum strength	Reduction of area
0.15 C., 0.38 Mn.....	Normal 850° C., water, tempered 300° C.	44 000	67 000	58
0.32 C.; 0.75 Mn.....	Normal 870° C., water, tempered 300° C.	135 000 53 000	175 000 85 000	11 56
0.60 C.; 0.42 Mn.....	Normal 800° C., oil, tempered 300° C.	124 000 75 000	202 000 123 000	29 38
1.02 C.; 0.42 Mn.....	Normal 820° C., oil, tempered 470° C.	183 000 110 000 183 000	203 000 162 000 218 000	17 15 16

In the case of the 0.32 per cent carbon steel, undoubtedly the yield point could have been kept higher than 124,000 pounds per square inch if the tempering temperature had been lower, for the reduction of area is larger than necessary.

**334.** In commercial practice case hardening is generally carried out by packing the steel pieces inside a cast-iron box together with carbonizing material, covering and sealing the box and heating for some hours at temperatures in the region of 900° to 1100° C. The carbonizing material is charred leather, or charcoal perhaps intermixed with barium carbonate or other chemicals. Carbon monoxide gas is either given off directly from the carbonizing material when heated, or is formed by the reaction of the carbonizing material with the oxygen of the occluded air. Carbon monoxide acts as a carrier of carbon from the carbonizing material into the surface of the steel. The chemical reaction, for which iron is a catalytic agent, is  $2\text{CO} = \text{CO}_2 + \text{C}$ . The carbon set free is taken into solid solution by the  $\gamma$  iron. The  $\text{CO}_2$  escaping from the surface of the iron reacts with the carbonizing material, again forming CO. The rapidity of the carbonization of the steel depends on the temperature and on the rapidity with which CO can be reformed from the carbonizing material. The carbon concentration possible in the surface of the steel depends upon the temperature, being fixed by the solubility of carbon in  $\gamma$  iron shown as the line *OPB* in Fig. 79. Unless the cementing is very rapid this limiting carbon concentration will not be reached in the surface of the steel, because diffusion through the solid solution is constantly carrying carbon from the surface layers inward. The diffusion speed increases with the temperature. The results as to the depth of case, and carbon concentration in the case, therefore, depend upon the balance of cementation speed and diffusion speed.

The rate of cementation depends upon the partial pressure of the CO gas in contact with the steel and on the relative amounts of CO and  $\text{CO}_2$  in the total gases. If there is too little CO, too much  $\text{CO}_2$ , present, the action becomes one of oxidation instead of carbonization. The necessary percentage of CO depends upon the temperature and upon the composition of the steel. It is generally considered that manganese, chromium, tungsten and molybdenum in the steel facilitate cementation, while nickel, titanium, silicon and aluminum make

it more difficult. For heat treatment after case-hardening see Section 303.

The recent studies in the theory and technique of case-hardening by Giollitti have resulted in the construction of case-hardening furnaces for the production of case-hardening work on a large scale, and have discovered methods for the exact control of the carbon content and depth of the case.

**335.** Below is given a table of the carbon contents of steel for various uses. Heat treatment is usually applied to all carbon tool steels and might be applied to all carbon steels. What those treatments are, or might be, may be seen from the preceding sections of this chapter.

#### TABLE OF CARBON CONTENTS AND USES OF CARBON STEELS

##### *Carbon 0.05 to 0.15 per cent*

Electrical sheet steel; boiler plate; rivets; bolts; stock for case hardening; nails; ship plate; forge work; sheet steel for tinning and galvanizing.

##### *Carbon 0.15 to 0.25 per cent*

General structural steel, for bridges, buildings, etc.; ship plate; rivets; forge work; cold rolled shafting.

##### *Carbon 0.25 to 0.40 per cent*

Machine parts generally; axles; shafts; connecting rods; piston rods; steel castings.

##### *Carbon 0.40 to 0.75 per cent*

Railroad rails (higher carbon in heavier rails); steel castings.

##### *Carbon 0.60 to 0.80 per cent*

Hammers; cutlery; wood working chisels, gouges, etc.; small lathe tools; small dies; cold sets; drills; taps; reamers; cold chisels.

##### *Carbon 0.80 to 1.00 per cent*

Carbon steel springs; ordinary lathe tools; large dies, drills and chisels; strongest steel wire and cable.

##### *Carbon 1.00 to 1.20 per cent*

Large lathe tools; dies and drills, large and carefully heat treated; axes; hatchets; knives.

##### *Carbon 1.20 to 1.50 per cent*

Saws; files; razors; jeweler's dies; balls and races of ball bearings (heat treated to remove free cementite).



## CHAPTER XVII

### CAST IRONS

**336.** Of the metals used in engineering structures, cast iron is by far the cheapest. Its physical properties are poor as compared with steels, since the strengths of cast irons rarely exceed 40,000 pounds per square inch, and, with the exception of the malleable variety, cast irons are brittle materials. From the blast furnace, the first stage of the reduction of ferrous metals from the ores, the product is pig iron. If pig irons are remelted and cast to desired shapes we have cast iron. If the pig irons are remelted and refined, by a comparatively expensive process, they are converted into steel, which may then be shaped by casting or hot working. Cast irons owe their cheapness to the cheapness of the material itself and to the cheapness of the method of shaping, casting. Due to the cheapness of the material, we can make up for its weakness and brittleness by using heavy sections, or comparatively large masses of metal. Between objects of the same material and shape shock resistance varies with the volume of the material, or the cube of the linear dimensions. Anyone who has tried to break up pig iron with a sledge-hammer realizes that cast iron in sections three to five inches square is hardly brittle to any ordinary attack.

In many of the uses of cast iron the factor of rigidity of the piece controls the design. Machine tools, such as lathes, may be cited as an example. To give to the lathe bed, the tool carriage, etc., sufficient rigidity to hold the cutting tool exactly to its work, would require considerable masses of either cast iron or steel, and still larger masses of other metals. The mass of metal required for rigidity of construction, so far as the material is concerned, varies inversely with the modulus of elasticity. Steel has the highest, and cast iron the next highest, modulus of elasticity among the engineering metals. To make a lathe bed rigid would take large masses even of steel, and would make the structure unnecessarily strong. If we make the lathe bed of cast iron, and use enough metal to make a rigid structure, we will have ample strength. In cases like this cast iron is very naturally chosen rather than steel because of the less cost.

**337.** Castings are most commonly made in sand molds. A "pattern" is first prepared of the shape and size of the desired object, with a certain allowance of oversize of the pattern for the anticipated shrinkage of the casting during cooling. Moist sand is then packed around the pattern. The sand is generally kept in two parts, held in the two halves of a "flask." Channels, called "sprues," are provided through what is to be the upper half of the mold ("cope") for the liquid metal to be poured into the mold. The mold is opened and the pattern removed, the interior of the mold smoothed up and the parts put together again. The casting is made by filling the hole in the mold with liquid metal, which in its freezing takes on the shape of the pattern.

If the sand is left moist, the hot metal causes the formation of steam which may interfere with the filling of the mold. The mold must, therefore, be "vented" to allow the escape, ahead of the entering metal, of this steam and of the air which might otherwise be trapped in the mold. If the casting is a complicated one, or of thin sections, or if the liquid metal is very hot as in steel casting, the steam formation must be avoided by drying out the mold before pouring. Since the so-called sand is really a mixture of sand and clay and loam, the organic matter in the loam may also be a source of gases when the sand is heated by the metal.

Where great quantities of similar castings are to be produced, machine molding is resorted to. The machine processes range from machinery for assisting in packing the sand around the patterns to almost completely automatic preparation of the molds.

Another device for cheapening the production of castings is the use of permanent metallic molds. This is more common in casting non-ferrous metals, especially aluminium alloys, but may be successfully applied to simple shapes of cast iron. The heavy cast-iron molds must be stripped from cast-iron castings as soon as these have become solid, or the mold itself will get too hot; and the stripping of the mold is also necessary to prevent interference with the shrinkage of the casting. Castings made in metallic molds are generally finer grained and better metal than those made in sand molds.

**338.** The greatest problem in the making of castings comes from the shrinkage of the metal. Usually a very considerable shrinkage

occurs during freezing. This makes necessary a continued supply of liquid metal to the casting while freezing is in progress. The need is met by the sprue and possibly by special "risers" or "sink heads" for feeding thin parts of the casting or parts remote from the sprue. The main part of shrinkage after freezing is the well-known phenomenon of contraction of solids by cooling or expansion by heating. With the majority of the non-ferrous metals this kind of solid shrinkage is all that occurs.

Cast irons and steels, however, present a much more complicated shrinkage action. There is first a shrinkage during freezing, large in steels and smaller in the cast irons. At the end of freezing, just as the metal becomes completely solid, there is in the gray cast irons a considerable expansion. From the end of freezing to the critical temperature there is regular metallic contraction. The decomposition of austenite at the critical temperature causes another expansion. Thence downward in the cooling the contraction is again regular. (See Section 364 and Fig. 150 for details.) Because of this succession of expansions and contractions, the danger of cooling stresses in steel and cast-iron castings is large. If the castings are uneven in thickness in different parts, the cooling will be uneven and one portion of the casting may try to expand at the same time the rest of the casting is shrinking. These possibilities control very largely the design of patterns.

**339.** The metal of a casting freezes first in the thin parts and last in the thick parts. It freezes more quickly on projecting corners than along surfaces, and last of all in reëntrant angles. It is very probable, therefore, that a sharp reëntrant angle will not be completely filled with metal, and that it will become a place of weakness in the casting. Filleting of all reëntrant angles is necessary.

**340.** Because it is taken as it comes from the ores without special purification, cast iron is a metal of complicated composition. It contains in addition to the iron considerable amounts of carbon, silicon, manganese, phosphorus and sulphur. Of these alloying elements, carbon and silicon must both always be considered as determining the properties; the actions of manganese, phosphorus and sulphur are usually secondary. In accounting for the nature of the internal structure of the cast-iron, the simplest method is to study first the pure iron-carbon alloy, and then bring out the modifications

affected in this by the other elements. The pure iron-carbon alloy of the cast-iron composition is of no industrial use because it is inferior in engineering properties to the cast irons of commerce, will not make sound castings and would be far more expensive.

The liquid cast iron is perhaps best thought of as a solution of  $\text{Fe}_3\text{C}$  in iron, just as salt may be dissolved in water. When this liquid solution freezes it forms two kinds of crystals,  $\gamma$  (C) (austenite with carbon in solid solution) and  $\text{Fe}_3\text{C}$  (cementite crystals). The freezing temperature of cast irons is about  $1130^\circ\text{C}$ . at the end of freezing, while the beginning of freezing is higher by an amount which varies with the composition of the metal. The beginning of freezing occurs at the temperature shown by the intersection on the equilibrium diagram, Fig. 79, of a vertical line representing the composition of the metal with the "liquidus" lines  $AD$  or  $DE$ . The end of freezing occurs on the lines  $BH$  or  $B'F$ . Crystalline  $\text{Fe}_3\text{C}$  is unstable at the freezing temperatures of cast irons and tends to decompose as fast as it forms. The time for decomposition of  $\text{Fe}_3\text{C}$  at  $1130^\circ\text{C}$ . is about 10 seconds. If the cooling of the casting during freezing is slow the  $\text{Fe}_3\text{C}$  decomposes, and the structure at the end of freezing is  $\gamma$  (C) + graphite. If the cooling during freezing is rapid, so that  $\text{Fe}_3\text{C}$  does not have time to decompose, the structure obtained at the end of freezing is  $\gamma$  (C) +  $\text{Fe}_3\text{C}$ . If graphite forms the casting is gray cast iron; if graphite does not form the casting is white cast iron. The theoretical relationships of white and gray cast iron have already been given in more detail in Sections 310 and 311.

**341.** The extreme range of composition of cast irons is from as low as two to as high as five per cent carbon; most cast irons come in the range from three to four and one-half per cent of carbon or carbon equivalents. If the composition of a cast iron lies to the left of the eutectic  $D$ , Fig. 79, the first crystals formed during freezing, or the "primary" crystals, are of  $\gamma$  (C). The end of freezing takes place with a filling in of the spaces between the primary crystals with a eutectic structure of  $\gamma$  (C) + G, or  $\gamma$  (C) +  $\text{Fe}_3\text{C}$ , depending on the rate of cooling. If the composition of the cast iron lies to the right of the eutectic point  $D$ , the primary crystals will be of G or  $\text{Fe}_3\text{C}$  and the eutectic structure surrounding the primary crystals correspondingly  $\gamma$  (C) + G, or  $\gamma$  (C) +  $\text{Fe}_3\text{C}$ , depending on the rate of cooling.

**342.** As the cast irons cool from the end of freezing to the critical temperatures, at about  $720^{\circ}\text{C.}$ , but one minor change of internal structure occurs. The austenite formed during freezing is at that time saturated with carbon in solid solution, approximately two per cent of carbon being held. As the temperature falls this austenite must throw out carbon because of the decreasing solubility shown by the lines *OPB* or *OPB'* in Fig. 79. The form taken by this carbon depends upon the carbon condition during freezing. If cementite was retained during freezing more cementite forms; if graphite was obtained during freezing more graphite forms. By this action the carbon concentration in the austenite is decreased to the point *O*, Fig. 79, when the temperature has reached the  $700^{\circ}\text{C.}$  region.

**343.** Theoretically, as the temperature goes below  $800^{\circ}\text{C.}$ , any graphite formed at higher temperatures ought to begin to recombine with the iron to form  $\text{Fe}_3\text{C}$ . Practically, such action if it occurs is negligible, at least in the case of coarsely crystalline graphite such as is formed during freezing. In the case of temper graphite this recombination may be perceptible. It is never important in the commercial handling of the cast irons. The theory of this action is given in Section 310.

**344.** In passing through the critical temperatures as they cool, cast-iron castings undergo a considerable change of structures. Austenite decomposes, yielding ferrite with carbon combined or free depending on the rate of cooling and the carbon form already present. It is a general fact of physical chemistry that a supercooled solution in decomposing tends to give metastable products first, unless the decomposition takes place in the presence of a stable product. Austenite decomposing in the presence of cementite at moderate rates of cooling, as in castings, yields pearlite, in which the carbon is present as  $\text{Fe}_3\text{C}$ . Austenite decomposing in the presence of graphite, under the same cooling conditions, yields temper graphite, or, as it is usually stated, the austenite changes to troostite. The  $\alpha$  iron of the troostite becomes ferrite, but the temper graphite may persist, or sorbite may form, depending somewhat on whether silicon is high or low.

Austenite being considerably more dense than troostite or pearlite, there is an appreciable expansion of the cast iron at the critical temperature. This expansion should be and is greater in gray irons

than in white ones, for troostite has greater volume than pearlite. After the changes at the critical temperature cast irons all cool without further changes and with regular metallic shrinkage down to ordinary temperatures.

**345.** We have now followed the freezing and cooling of the cast irons, bringing out that if they freeze to white cast iron they stay white during cooling, with the formation of secondary  $\text{Fe}_3\text{C}$  from austenite decomposition both preceding and at the critical temperature; if they freeze to gray cast iron they stay gray during cooling, additional (temper) graphite forming from austenite decomposition before and at the critical temperature. It is evident that **the determination of the metal as white or gray cast iron occurs during freezing.** The controlling factor is generally the rate of cooling during freezing, which may or may not allow time for the decomposition of  $\text{Fe}_3\text{C}$  preceding and at the end of freezing.

**346.** Of the factors determining the rate of cooling during freezing, we may put first the composition of the metal. Increase of the carbon or silicon content of a cast iron causes an increase of the latent heat of freezing and correspondingly decreases the rate of cooling of a casting during its freezing. Therefore cast irons of high carbon, or high silicon, or both, are almost certain to be gray irons; while cast irons of low carbon and low silicon tend to be white.

**347.** A minor composition factor is the amount of sulphur in the cast iron. The method by which the sulphur acts has never been explained. The effect is that even small amounts of sulphur may markedly increase the amount of cementite formed and the persistence of this cementite after its formation. The influence of a given weight percentage of sulphur sometimes balances about fifty times as great a weight of silicon acting in the opposite direction. For further detail on the action of sulphur see Section 366.

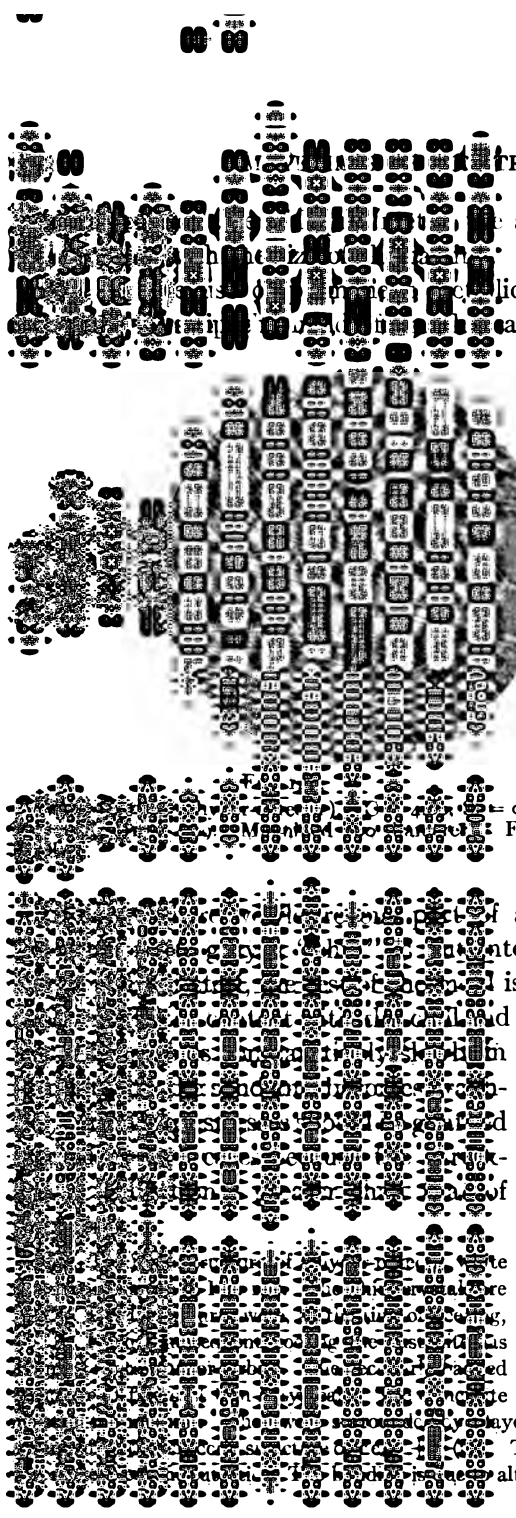
It may be pointed out that the work of Dr. Stead on the iron-carbon-sulphur alloys shows that some sulphide of iron, commonly considered to be  $\text{FeS}$ , is present in white cast iron mechanically included in both primary and eutectic cementite crystals. This could not be possible unless solid sulphide crystallites were present in the liquid cast iron before the beginning of formation of even the primary cementite. We have no explanation of this indubitable fact. In the iron-sulphur system the freezing point of pure  $\text{FeS}$  is supposed to be about  $1200^\circ \text{C}$ ., and a eutectic between  $\text{Fe}$  and  $\text{FeS}$  has a temperature under  $1000^\circ \text{C}$ . It may also be pointed out that the iron-carbon-sulphur alloys do not show any traces of a eutectic of three solid phases, corresponding to the eutectic found, for

instance, in iron-carbon-phosphorus alloys. This is consistent with the finding of the sulphide mechanically included in cementite crystals, indicating that the sulphide was solid before the beginning of carbide solidification. Stead reports also that chemical analysis shows a little sulphur in solid solution in  $\text{Fe}_3\text{C}$ . This might account for the increased stability of the  $\text{Fe}_3\text{C}$ ; but this might also be due to greater size of the  $\text{Fe}_3\text{C}$  crystals.

**348.** Elements other than carbon, silicon and sulphur seem to have little effect during the freezing of cast iron in determining whether the iron becomes white or gray. Manganese is often credited with a tendency to make white irons; experiment does not confirm this. If a cast iron does become white during freezing the presence of manganese in any appreciable amounts makes it hard to decompose the  $\text{Fe}_3\text{C}$ . Manganese forms a stable carbide  $\text{Mn}_3\text{C}$ , taking carbon away from the iron. This  $\text{Mn}_3\text{C}$  forms a solid solution with  $\text{Fe}_3\text{C}$ , making the apparent cementite crystals of a manganiferous cast iron comparatively stable. Vanadium has a similar action. Phosphorus occurs in the cast irons as  $\text{Fe}_3\text{P}$ . The formation of this compound takes up some of the iron so that the carbon concentration in the remaining iron is increased. The result is an almost negligible tendency to a greater graphite formation in the presence of phosphorus.

**349.** The external factors determining the rate of cooling during freezing, and consequently the formation of white or gray cast iron, are the temperature of pouring, the shape and size of the casting and the material of the mold. The liquid metal in the mold cools by a time temperature relation such as shown in Fig. 114. If the metal is poured "cold," the outer part of the casting cools sharply, in heating up the mold, and the center cools very slowly during freezing. If the metal is poured "hot," the mold will be thoroughly heated long before freezing, and the piece will be cooling rapidly throughout during freezing. At the proper pouring temperature we avoid both of these extremes, the piece cooling throughout at a moderate rate during freezing.

**350.** The amount of heat to be given up by the casting to the mold depends on the mass of the casting, while the ability to get rid of heat depends upon the area of the surface of the casting. Therefore shape and size of castings influence the cooling rate during freezing. The relations are the same as those discussed in the quenching of steel in Section 297. As a result of this effect of size one finds that if a number of castings of different sizes and similar shapes are poured



# STRUCTURE

amount of graphite formed in chill molds on cast iron, it is castings as soon as the castings are solid, but while their temperature is still close to the freezing point. The cementite formed during freezing of the chilled casting thus has an opportunity to decompose in the solid casting. The graphite takes the form of temper graphite. The same metal cast in sand molds would form graphite in large crystals during freezing.

a casting to be white and into the mold. The chill is a is sand. Part of the casting and remains white. The rest of

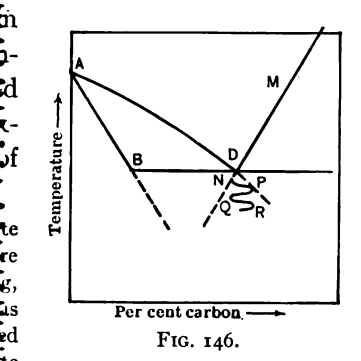


FIG. 146.

layer of austenite. Next there formed The last part to freeze was a similar, alternate overproduction of  $Fe_3C$  and



ite crystals began to form during the melt during continued cooling curve  $MDN$ . At the point  $D$  the liquid, crosses the liquidus curve should be completed at the point  $E$  in a eutectic structure. Actually the formation continues until the remaining some point  $N$ . The farther the tendency to formation of  $\gamma$  ( $C$ ). At  $N$  composition of the melt shifting to the an overproduction of  $Fe_3C$  in a in a eutectic structure. Instead actually oscillates between over- Similar banding in eutectic struc- metals.

outside than inside, the natural the central parts and correspond- on the edge of being either white the nature of the metal formed. the same chemical composition. white, while a tension stress helps to



FIG. 147.

structure of white Swedish iron. the black patches represent spherical eutectic formation shown in transverse section. From Hatfield, "Cast Metals." About  $\frac{1}{4}$  natural size.

iron scattered through the white. and would continue until the whole stop to it.

equilibrium relations of cast iron-carbon alloys, Fig. 79,

as if it were true for the cast irons. The other chemical elements of the commercial cast irons, silicon, sulphur, phosphorus and manganese, modify the iron-carbon relationships. The most important of these elements is silicon. The effects of the elements other than carbon are noticed first in the composition of the eutectic, point *D* of Fig. 79. Adding silicon or phosphorus decreases the carbon content of the eutectic. Each per cent of silicon lowers the carbon content of the eutectic by 0.27 per cent; each per cent of phosphorus causes a similar change of 0.34 per cent. The effects of sulphur and manganese on the eutectic composition are insignificant.

**356.** In the pure iron-carbon alloys the carbon content of the  $\gamma(C) + G$  eutectic is 4.25 per cent. With the addition of silicon and phosphorus, the carbon content of the  $\gamma(C) + G$  eutectic in the commercial cast irons is  $4.25 - 0.27 \times \text{per cent Si} - 0.34 \times \text{per cent P}$ . This percentage of carbon at the eutectic point is known among foundrymen as the "saturation" value. When a cast iron of carbon content greater than saturation value is made in the blast furnace and run out to make pig iron, the freezing of such iron is peculiar. In the comparatively slow cooling of the pigs the primary crystals forming from the liquid metal are in this case graphite crystals. Graphite has a density of about 2.25, and the liquid cast iron has a density of about 6.5. This difference of densities is so large that the bulk of the primary graphite crystals float up to the surface of the liquid cast iron. At the end of the freezing both  $\gamma(C)$  and graphite crystals form simultaneously at the eutectic temperature. Any graphite remaining in the metal after the iron crystals begin to form is caught and held. The result is the retention in the solid metal of a total carbon content slightly in excess of the eutectic or saturation percentage. The remainder of the carbon has escaped to the upper surface of the casting, and is there present as a layer of flaky graphite crystals known technically as "kish." In the handling of the pig iron the kish rubs off.

**357.** Of far more importance in the commercial cast irons than their effects on the carbon content of the eutectic, are the effects in the solid metal of the chemical elements other than carbon. The most important of these elements is again silicon. The silicon goes almost entirely into solid solution in the austenite. In white cast

irons some silicon is found in the apparent  $\text{Fe}_3\text{C}$ , probably by the formation of  $\text{Fe}_3\text{Si}$  going into solution in  $\text{Fe}_3\text{C}$ . As silicon goes into solid solution in austenite, forming the multiple solution  $\gamma$  (C, Si), the limiting carbon content of the solution is lowered by the presence of silicon. In the cast irons this solid solution is always saturated, and the effect of silicon addition is therefore to throw carbon out of the solid solution form, forcing it to become either  $\text{Fe}_3\text{C}$  or G. The stages in which the displacement of carbon by silicon proceeds are illustrated in Fig. 148. The line  $MN$  is the boundary of the multiple solid solution  $\gamma$  (C, Si) at the freezing

temperature of approximately  $1140^\circ\text{C}$ . The line  $PQ$  is the similar boundary when the temperature has dropped to the critical, at a little above  $700^\circ\text{C}$ . (Neither line is actually quite straight as drawn.)

In using this diagram we must always remember that silicon displaces carbon in the solid solution.

Suppose an iron-carbon-silicon al-

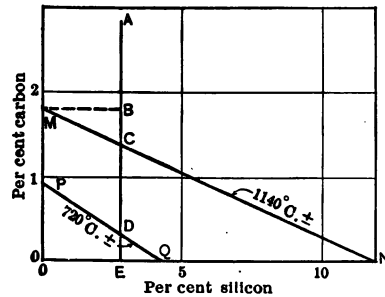


FIG. 148.

loy of the composition of point A, Fig. 148, is cast. If there were no silicon this would become at freezing austenite with carbon in solid solution to the amount  $EB$ , and carbon outside the solid solution would be  $BA$ . Because of the presence of silicon the distribution of carbon at freezing is that the amount  $EC$  is in solid solution and  $CA$  is outside the solution. During the cooling from freezing to the critical temperature the saturation limit of the solid solution decreases, further carbon to the amount  $CD$  being thrown out of solution. The  $\gamma$  (C, Si) decomposing at the critical temperature would in this case contain all of the silicon and only the amount  $ED$  of the total carbon. This amount  $ED$  of the total carbon will change generally at the critical temperature to pearlite or sorbite. In gray cast irons this carbon constitutes the combined carbon of the cast iron. The carbon amount  $DA$  which at various stages of the cooling has been rejected by the solid solution will probably be in the form taken by carbon during freezing, either  $\text{Fe}_3\text{C}$  or graphite. Because silicon increases the latent heat of freezing, we will find that with high silicon

content (generally with more than one to two per cent) the carbon rejected from solid solution becomes graphite.

**358.** Below the critical temperature silicon is soluble in  $\alpha$  iron to the extent of about two per cent. The silico-ferrite is stronger and harder than pure  $\alpha$  ferrite by about thirty per cent at two per cent silicon, but the ductility at the same time is decreased by half. Silicon between two and four per cent forms a pearlitic structure of  $\alpha$  (Si = 2) and  $\text{Fe}_2\text{Si}$ , strong (about 80,000 pounds per square inch) but brittle. Beyond four per cent of silicon free crystals of  $\text{Fe}_2\text{Si}$  occur with the silico-pearlite, and brittleness increases while strength decreases. These actions are similar to the changes of properties with carbon content in steels, shown in Figs. 97 and 98 and their discussion.

**359.** By combining the approximate numerical values of the various displacements of carbon by silicon, also making allowance for the effect of silicon on the latent heat and the corresponding effect on graphite formation during freezing, we may summarize with sufficient accuracy the silicon effect as a carbon displacement equal in amount to one-half of the silicon content of the cast iron. We may use the equilibrium diagram of Fig. 79 fairly well for the cast iron if, instead of entering the diagram by the abscissa of actual total carbon, we use an **equivalent total carbon** value equal to the actual total carbon plus one-half the silicon.

**360.** Silicon in the commercial cast irons has one important effect which makes it always necessary. It cleans and largely deoxidizes the liquid metal, and so makes sound castings possible. Still further deoxidation may be had with vanadium or titanium; but silicon does most of the work, and does it far more cheaply than any other agent.

**361.** In the iron-phosphorus alloys, the relations of iron to phosphorus are very similar to the relations of iron to carbon in their alloys. Phosphorus not going into solid solution  $\gamma$  (P) during freezing forms  $\text{Fe}_3\text{P}$ , which is very similar in properties to  $\text{Fe}_3\text{C}$ .  $\text{Fe}_3\text{P}$  probably does not, however, form solid solutions with  $\text{Fe}_3\text{C}$ . In the iron-carbon-phosphorus alloys carbon displaces phosphorus from solid solution, just as silicon displaces carbon in the iron-carbon-silicon alloys. If the austenite solid solution in the cast irons were not saturated, it would contain silicon, carbon and phosphorus,  $\gamma$  (C, Si, P). Since the solution is saturated, and both carbon and silicon displace

is found even from the  
ing phosphorus retain the  
ssible. The first crystals

Eutectic of  $\gamma$  (C Si)  
+ G +  $\text{Fe}_3\text{P}$



FIG. 149.

st iron with high phosphorus.  
Si = 3.43, P = 1.31, Mn = 0.49.  
200 diameters. From Guillet.

primary crystals were graphite.  
P is plainly visible. (The  
and pearlite at the critical  
appearance of the micro-

$\text{Fe}_3\text{C}$  to increase hardness  
strength. Since the cast  
are not spoiled by the pres-

considerable amount of iron  
of carbon in the re-  
effect so far as the carbon  
the total carbon by an  
present.

in the cast irons is due  
 $\text{Fe}_3\text{C}$ , which goes into solid  
Manganese may go  
iron. The austenite of the

cast irons at temperatures above the critical contains more or less manganese in solution. At lower temperatures this manganese is found in solid solution in the  $\alpha$  iron. It does not affect the equivalent carbon values perceptibly.

**363.** The effect of sulphur in cast irons has been discussed in some detail in Section 347. Vanadium may be used as a cleaning agent for oxides; in case any vanadium remains it has actions similar to manganese. Titanium may be used as a cleaning agent for nitrides and oxides; any remaining after the cleaning action behaves like silicon. Aluminum may be used as a cleaning agent for oxides; any excess acts much like silicon. When elements are used as cleaning agents in cast irons the amounts used are only a few tenths of one per cent, and the amounts remaining in the castings are still smaller. The beneficial effect of getting rid of oxides and nitrides is quite perceptible in the engineering properties of the castings.

**364.** It has been pointed out that the greatest problem to be solved in the making of castings comes from the shrinkage of the metal during cooling. The curves of the shrinkage of various metals versus

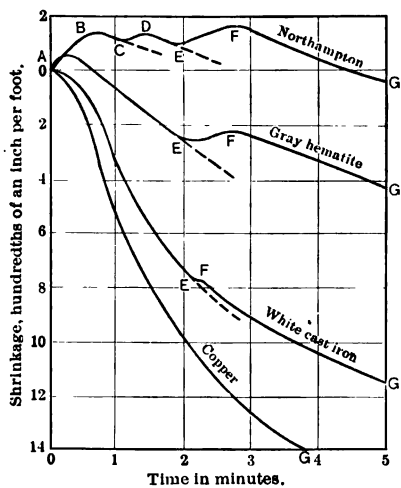


FIG. 150.

time after beginning of solidification are shown in Fig. 150, from the work of Professor Turner. The curves as drawn are not carried down to the completely cold condition, but the shrinkage beyond that shown in the figure would be regular solid shrinkage for all the metals concerned. The temperatures at the points G are still three to four hundred degrees Centigrade. The lowest curve is that for copper. It has no break in its regular shrinkage from the freezing point to cold metal. Similar curves are given by aluminum,

lead, tin, zinc and the majority of the non-ferrous alloys. The curve for white cast iron shows one period of expansion, at *EF*, which was shown to be due to the change of austenite to pearlite. Steels show this same

change for the same reason. A similar single break is found in some of the brasses and bronzes due to allotropic changes. The curve for gray hematite is fairly typical of gray cast irons of low phosphorus content. At the end of freezing there is an appreciable expansion due to graphite formation. This is succeeded by regular shrinkage to the critical temperature where there is a second considerable expansion at *EF*. The curve marked Northampton is for an iron high in silicon, carbon and phosphorus. The expansion from *A* to *B* is due to graphite formation during the main freezing period. The expansion from *C* to *D* is due to graphite formation in the  $\gamma$  (C, Si) + G + Fe<sub>3</sub>P eutectic. Below *D* the metal is entirely solid. The expansion from *E* to *F* is due to change of  $\gamma$  to  $\alpha$  iron at the critical temperature and accompanying changes. It will be seen that due to the graphite formation during freezing and the greater volume change at the critical temperature, the net shrinkage from size of pattern to size of cold casting is much less for the gray irons than for the white irons and steels. At temperatures above the critical the cast iron is quite weak and easily broken in the molds by shrinkage stresses. Gray irons are much easier to cast than white irons because their shrinkage in this critical region is small; and a phosphoric gray iron is the easiest of all cast irons to cast.

**365.** It will be evident from the above discussion that control of shrinkage depends mostly upon control of the graphite formed in the cast irons and the temperature at which the graphite forms. The foundryman obtains this control by varying the percentages of carbon, silicon and phosphorus in the metal. The amount of shrinkage must be made to suit the kind and size of castings being made. In general increase of graphite weakens a cast iron, so that the graphite amount should be kept to the minimum which will give a sufficient reduction of shrinkage.

**366.** The relation between the amounts of carbon, silicon and phosphorus in a cast iron and the amount of graphite formed, is indicated in Fig. 151. The abscissa of this figure is the equivalent total carbon, which is the summation of actual total carbon plus 0.5 silicon plus 0.1 phosphorus. The points plotted represent data from both pure iron-carbon alloys and a large number of commercial pig irons. It will be seen that in general the amount of graphite formed comes

within the band between the two lines *ABC* and *DEF*. Up to an equivalent carbon content of 3.5 per cent the latent heat of the cast irons is small enough so that  $\text{Fe}_3\text{C}$  does not decompose under commercial casting conditions. Beyond four per cent of equivalent carbon the latent heats of cast irons are large enough so that graphite always forms in sand castings and may even form in chilled castings. With slow cooling during freezing, as with large castings in sand, the graphite formed will approach the amounts indicated along the

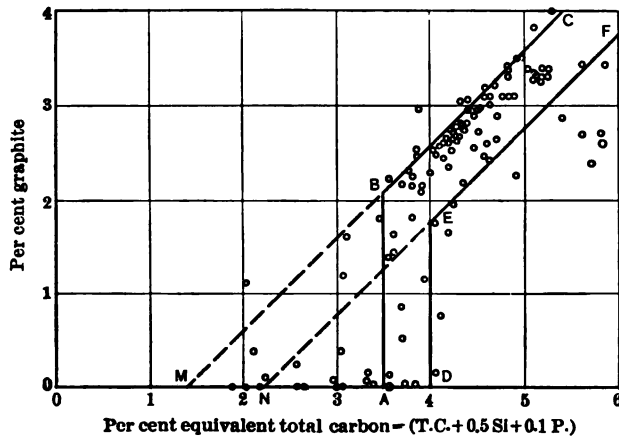


FIG. 151.

line *ABC*. With the quicker cooling of small castings in sand or of large castings in chills, the amount of graphite formed approaches the limits of the line *DEF*. With small castings in chills graphite formation can be suppressed in practically any cast iron; while with artificially slow cooling during freezing graphite may be formed in considerable amounts with equivalent carbon contents below 3.5 per cent, realizing the line *MB* as a graphite limit. The effect of sulphur is erratic. If the equivalent total carbon is such that the castings might be white, high sulphur will make them strongly so; if the equivalent total carbon is high, so that the castings would be normally gray, sulphur has little effect.

**367.** High phosphorus (up to 1.5 per cent) in cast irons gives one especial advantage, which may more than counterbalance the slight loss of strength and increase of brittleness from the presence of considerable amounts of the  $\text{Fe}_3\text{P}$  resulting. Cast irons without phosphorus



becomes solid above  $1100^{\circ}\text{C}$ .; cast irons with high phosphorus remain partially liquid down to below  $950^{\circ}\text{C}$ . With lower freezing point, or greater fluidity at the same temperatures while liquid, and less shrinkage than other cast irons, the high phosphorus irons can be poured successfully into thinner and more complicated castings.

**368.** The engineering properties of the cast irons are fixed by (1) the amount and condition of the carbon, (2) the amount of the silicon, and (3), in a minor way, by the amounts of phosphorus, manganese and sulphur. The carbon condition may be  $\text{Fe}_3\text{C}$ , in large primary crystals, or in eutectic structures similar to pearlite, or both; or it may be graphite, in large primary plate crystals, in small eutectic crystals, or exceedingly minute temper graphite.  $\text{Fe}_3\text{C}$  in any form hardens the cast iron, since  $\text{Fe}_3\text{C}$  is harder than glass.  $\text{Fe}_3\text{C}$  in pearlitic structures strengthens the iron and also hardens it; free  $\text{Fe}_3\text{C}$  crystals increase hardness and do not increase strength.  $\text{Fe}_3\text{P}$  acts like  $\text{Fe}_3\text{C}$ . Graphite in any form softens the iron, and also weakens it. The weakening by graphite is least by temper graphite, and greatest by large primary graphite crystals. Silicon up to two per cent goes into solid solution as silico-ferrite, hardening and strengthening the iron. Silicon beyond this amount does not greatly increase the strength; indeed may lower it, while it does increase the hardness, due to the formation of  $\text{Fe}_2\text{Si}$ .

The various factors in strength and hardness cannot be made to work together, because of the effects of silicon on the condition of the carbon. White cast irons are so hard that they can be machined only with difficulty. The carbon must be made largely graphitic in order to make the cast iron soft enough to machine well; and this advantage is secured largely at the cost of the strength of the metal. The carbon must be made largely graphitic also for another reason, to reduce the shrinkage to manageable values.

**369.** The usual compositions and properties of cast irons are tabulated on the following page.

The strongest cast iron which is also soft enough to machine well is the light gray. When a light gray cast iron is made by melting in the cupola a full gray pig iron with 20 to 30 per cent of its weight of scrap steel, to dilute the total carbon and silicon down to the values for light gray iron, the product is called "semi-steel." The same name is also sometimes, dishonestly or mistakenly, applied to malleable cast iron.

Variety.	Chemical analysis.							Strength.	Ductility.	Remarks.	
	Total Carbon.	Silicon.	Phosphorus.	Sulphur.	Manganese.	Equivalent total carbon.	Graphite.				Combined carbon.
Ordinary gray .....	3.5±	2	Usually under 0.5 Sometimes 1.5	Usually under 0.2	Usually under 1.0	4.5±	3.0	0.5±	15,000 25,000	0	Readily machined.
Light gray.	3.5 3.0	1.5 1.0				3.9±	2.0+	1.0—	25,000 35,000	0	Harder to machine.
Half white.	3.0 2.5	1.0				3.7±	small	2.0+	25,000 35,000	0	Difficult to machine.
White.....	3.0 2.0	0.8 0.4				3.4—	0	All	About 45,000	0	Difficult to machine.
Malleable...	2.5 3.0	1.0 0.3	0.25 0.10	0.10 0.02	0.70 0.20	3.0±	?	?	35,000 50,000	8 to 2%	Heat treated to change bulk of Fe <sub>3</sub> C to temper G. Soft and tough.

The bulk of commercial cast irons, when used with no other operation after casting than machining, are gray irons. The full gray is preferred except when the higher strength of the light gray controls the choice. A lathe bed, for instance, would be light gray, partly for strength, partly because a light gray is harder than full gray iron and will wear longer. It must be remembered also that the amount of graphite formed in a casting, and hence the properties of the metal, depends on the size of the casting as well as upon the composition of the cast iron used.

White cast irons find a few special uses on account of their hardness, which is retained unimpaired up to a red heat. Such uses are for rolls in steel and paper mills. Most white cast iron is heat treated to convert it into malleable cast iron.

Tension test curves of cast irons are shown in Fig. 152. The curves are all drawn in once to a large deformation scaling to show the relative actions at the breaking loads, and are then repeated separately on a magnified scale to show relative stiffnesses and the rather ill-defined yield point of the malleable.

**370.** When a cast iron is melted in a foundry cupola slight amounts of both silicon and carbon are burnt out of the iron. Continued remelting of the same iron will reduce its equivalent total carbon so that it will change from gray to white. Allowance for this loss of

silicon and carbon in melting may be made by adding to the charge a compensating amount of high-silicon pig iron, or by adding ferro-

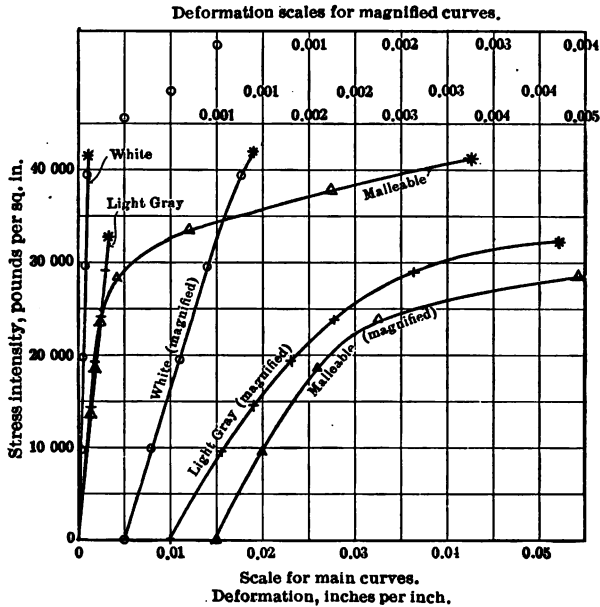


FIG. 152.

Comparative tension test curves of white, light gray, and malleable cast irons.

silicon in the ladle after melting. A foundry which remelts its sprues and defective castings, and neglects these precautions, will sooner or later have its castings come hard and unsound.

371. Cast irons will respond to annealing or hardening and tempering heat treatments much as steels do. Annealing for removal of cooling stresses is not an uncommon treatment for complicated castings. It is especially desirable for such an article as a block cylinder casting, or a piston, for an automobile engine. The annealing, consisting of slow heating to dull red and slow cooling, should best come between the roughing and finishing operations in the machining of the cylinders or pistons. Such annealing will anticipate any permanent deformations by heating of the cylinders or pistons in operation of the engine.

White cast irons do not need any further hardening by heat treatment. It can be done, but the cooling stresses will probably be so

severe as to break the casting to pieces during quenching. Light gray cast irons may be made as hard as tool steel by the same treatment as given the steel, but will have high cooling stresses. Full gray cast irons may or may not respond to heat treatment like steels; their action depends on the amount of silicon and the amount and size of graphite crystals. A cast iron which when heated above the critical temperature became  $\gamma$  (Si) + G, with no carbon in solid solution in the austenite, would not harden during quenching, because silicon does not slow up the decomposition of  $\gamma$  (Si) enough to make it controllable. If the cast iron, heated above the critical temperature, consists of  $\gamma$  (C, Si) + G, the response to heat treatment depends upon the amount of carbon in the  $\gamma$  (C, Si). The presence of graphite crystals in the gray cast irons, or of free  $\text{Fe}_3\text{C}$  crystals in the white ones, makes cooling stresses during quenching or tempering operations very likely to break the metal. The hardening of cast irons by heat treatment has practically no commercial application.

**372.** One peculiar action of gray cast irons may become serious in commercial use of the iron. Any cast iron containing silicon and graphite, when heated red hot and cooled back, grows a little. Repeated heating and cooling causes a cumulative growth, which may ultimately amount to a volume increase of 40 to 60 per cent over the original size. Such action is found in cast-iron grate bars and other parts of furnaces. In case one side of the metal is hotter than the other during the heating, the greater growth of the hot side warps the piece. The weight of the metal increases during the growth, and the strength decreases until it practically vanishes.

The cause of the growth is an oxidation. Oxygen from the air or furnace gases penetrates the metal along the graphite crystals, oxidizing the graphite, the combined carbon, the silicon and ultimately even the iron. In the absence of graphite crystals the gases cannot penetrate below the surface. The loss of strength and growth in weight and volume seem to be due to formation of  $\text{SiO}_2$  within the metal.

**373.** A heat treatment of cast irons which is of great commercial importance is that which changes white into malleable. Malleable cast iron has sufficient ductility to give it very considerable shock resistance, while its strength approaches that of wrought iron or a

low carbon steel. It is considerably cheaper than wrought iron or steel so that it competes commercially with them as a structural material. The draft and coupling gear of freight cars, or the greater part of agricultural machinery, or hardware and pipe fittings, are examples of the use of malleable cast iron. (Malleable cast iron finds wide commercial use only in America. In England the name "malleable" refers to the material known in America as wrought iron.)

**374.** The white cast iron to be converted into malleable should have its chemical specifications of phosphorus, sulphur and manganese similar to those of steels. Phosphorus is commonly limited by specification to a maximum of 0.25 per cent, but might better be down to 0.10 per cent; sulphur is allowed up to 0.10 per cent, but should be below 0.06 per cent; manganese is allowed up to 0.70 per cent, but for the best results should not exceed 0.30 per cent. The total carbon may vary from 3.0 per cent in small castings to 2.5 per cent in large castings, while the silicon varies from 1.0 per cent for small castings to as low as 0.3 per cent for castings several inches thick in solid sections. The intention in choice of carbon and silicon contents is to make the casting come out white cast iron, just as near as may be to the formation of graphite during casting. High manganese must be avoided because it makes the manganese containing cementite hard to break down during malleableizing. Phosphorus and sulphur must be kept low for the same reasons as in steels. The castings are made with a shrinkage allowance the same as for gray cast iron, not for white cast iron, because the expansion due to formation of temper carbon during malleableizing causes a volume increase about equal to the difference of shrinkage between white and gray cast irons.

**375.** The malleableizing process is essentially one of decomposition of cementite into iron and temper graphite by continued heating at temperatures in the region of  $800^{\circ}$  to  $1100^{\circ}$  C., where  $\text{Fe}_3\text{C}$  is unstable. At temperatures near  $1100^{\circ}$  C. the breakdown of the  $\text{Fe}_3\text{C}$  may be started in a few minutes; near  $800^{\circ}$  C. it may take from one to three days. The product obtained by malleableizing at high temperatures is coarser grained and weaker than from lower temperatures. Commercially, therefore, the temperatures used for malleableizing are a little above  $800^{\circ}$  C.

The castings of white iron are packed into large cast-iron boxes, being completely surrounded by packing material, which is commonly mill scale (black iron oxide,  $\text{Fe}_3\text{O}_4$ ). The packing material may be fire clay, sand, charcoal or coke. The main purpose of the packing is to support the castings while they are heated, to prevent sagging out of shape by their own weight. The oxygen given off from the mill scale, or oxygen of air occluded in any of the packings, plays a secondary part in the malleableizing process. The boxes of castings and packing are covered over and sealed. They are heated for a day or so in coming up to temperature and for two or more days at constant temperature, and are then cooled for several days in the furnace. The progress of the malleableizing is followed by pulling out special test pieces at intervals and examining the fracture of these pieces.

At the temperature of malleableizing the internal structure of the castings consists of  $\gamma$  (C, Si) +  $\text{Fe}_3\text{C}$ . To be more precise, both silicon and manganese are present to some extent in the cementite and there is manganese in the austenite. Phosphorus will be in the form of  $\text{Fe}_3\text{P}$ , and sulphur as  $\text{MnS}$ . With low manganese the cementite will break down spontaneously in a few days of exposure at just above  $800^\circ\text{C}$ . This would happen throughout the casting, no matter how thick. The structure would become  $\gamma$  (C, Si) + temper graphite. Because of the silicon in the austenite the amount of carbon in solid solution cannot be very large. (Approximately the possible carbon in  $\gamma$  (C, Si) at  $800^\circ\text{C}$ . equals  $1.0 - 0.25\text{ Si}$ . With 0.8 per cent silicon this means a carbon in austenite of  $1.0 - 0.2 = 0.8$  per cent.)

**376.** In the commercial process the decomposition of the  $\text{Fe}_3\text{C}$  is accelerated by the action of the oxidizing atmosphere. The oxidizing gases penetrate the surface of the castings to a depth of about  $\frac{1}{8}$  inch. To that depth they ultimately remove all of the carbon, both that in the  $\text{Fe}_3\text{C}$  and that in  $\gamma$  (C, Si). Then on the boundary between the decarbonized surface and the core of the casting the decarbonized austenite starts to react with  $\text{Fe}_3\text{C}$ , to restore its own carbon content to saturation. This reaction becomes the trigger starting the general decomposition of the  $\text{Fe}_3\text{C}$ . The decomposition of the  $\text{Fe}_3\text{C}$  then proceeds rapidly inward, each crystal in its decomposition starting

the next one by contact action. The decomposition of the  $\text{Fe}_3\text{C}$  spreads far more rapidly than diffusion of carbon through austenite solid solution. Because by the oxidation the breakdown of the  $\text{Fe}_3\text{C}$  is started at the surface, we may find that a malleable piece, where the process was stopped too soon, contains a white core. Such cores are more likely to be due to segregation of manganese during casting, making the manganese containing cementite in the center of the casting resistant to the heat treatment. As a result of the series of actions a cross-section of the final product will show an outer rim of practically zero carbon content to the depth to which the oxidizing gases penetrated; thence black to the center, though there may be a brittle white or grayish core. The black fracture is due to temper graphite. Inside the white outer rim the carbon content jumps suddenly, through a narrow transition layer where diffusion through the austenite had been pulling carbon out toward the surface, to the total carbon value of the original white cast iron; but the carbon is now mostly in the form of temper graphite. The black fracture gives to the malleable cast iron the commercial name of "black-heart" malleable.

**377.** During the slow cooling in the boxes at the end of the malleableizing process the  $\gamma$  (C, Si) solid solution decomposes at about  $700^\circ\text{C}$ . The products of the change are  $\alpha$  (Si) +  $\text{Fe}_3\text{C}$  + temper graphite, or more definitely silico-ferrite and mixtures of this with the other crystals in pearlite or sorbite or troostite. Inspection of test pieces quenched from before and after the decomposition of the austenite will often show a considerable increase of temper graphite during the change. This is probably due to a contact effect of the previous temper graphite upon the decomposing solid solution, determining the products of decomposition of the solution, so that troostite forms rather than the expected pearlite. (See section 344.) The properties of the finished malleable cast iron, given in the table in Section 369, depend upon the amounts of silico-ferrite, pearlite and temper graphite in the structure. This structure is shown in Fig. 135.

**378.** It is often assumed that the white outer rim of the black-heart malleable is stronger or more ductile than the inner parts. It is probably, on account of its carbonless condition, weaker but more

ductile than the inner parts which are hardened by pearlite. A white core in a piece of malleable is both weak and brittle.

**379.** While heat treatment of black-heart malleable is not commercially used, a few experiments by the author have shown that it may be worth investigation. Half-inch round bars of commercial malleable were heated to between  $750^{\circ}$  and  $800^{\circ}$  C. and oil quenched, without tempering. They gave a strength in tension of 75,000 pounds per square inch, but were brittle. With a slight tempering they might have retained most of this strength while gaining enough ductility to give shock resistance. The fractures showed that an appreciable amount of temper graphite had been reabsorbed during the reheating. The same material untreated had a strength of 41,000 pounds per square inch with about 4 per cent ductility.

**380.** Black-heart malleable is used almost entirely in America. Throughout Europe the malleable cast iron is generally "white-heart." The process is similar to the American process, but the temperatures are considerably higher and the times of heating longer, so that the oxidation of carbon goes entirely through the metal. Diffusion of carbon through the solid solution helps to pull carbon from the inner parts to the surface. White-heart malleable can hardly be made in sections thicker than  $\frac{3}{8}$  inch, because it takes too long to get the carbon out. The process is more costly than the black-heart process and the product is certainly no better in engineering properties.



## CHAPTER XVIII

### ALLOY STEELS

**381.** Alloy steel is the name given to any steel to which considerable amounts of one or more chemical elements other than iron and carbon have been intentionally added. The alloying is done to get some better set of properties than can be had in carbon steels, as, for instance, higher strength with the same ductility, or increased resistance to shock or fatigue, or to make a steel more readily or reliably controlled in its properties by heat treatment.

The carbon steels used commercially rarely have carbon in excess of one per cent. The same restriction as to carbon content is to be understood in the following discussion of alloy steels.

**382.** We will study first the normal structures, which are the result of the usual commercial processes of hot working for shaping, followed by cooling in air from the red heat to ordinary temperatures. Three different structural types, each with its peculiar properties, are found as normal structures in the alloy steels.

(1) **Pearlitic.** This is the normal structure of the commercial carbon steels. It consists of mixtures of crystals of  $\alpha$  iron with patches of sorbite or pearlite. The amount of pearlite is proportional to the carbon content, reaching 100 per cent at about 0.9 per cent carbon. The alloying element in the pearlitic steels is usually in solid solution in the ferrite. Occasionally there is also a carbide of the alloying element similar to  $\text{Fe}_3\text{C}$  and crystallizing in solid solution in it.

Each increase of 0.1 per cent of carbon in the pearlitic structure changes the properties by raising the maximum strength about 10,000 pounds per square inch, by raising the yield point about 6000 pounds, and by lowering the ductility and the reduction of area 3.5 and 7.0 per cent respectively. The ratio of yield point to maximum strength is about 0.6. Pearlitic steels are easily magnetized to considerable flux density, but do not retain their magnetism.

(2) **Martensitic.** This is the structure found in the fully hardened high-carbon tool steels. It is due to  $\beta$  iron, non-crystalline (see Sections 317 to 321). The carbon is mostly in solid solution. Barring cooling stresses and check cracks the strength is very high — pure martensite without cooling stresses might have a strength near 400,000 pounds per square inch. The yield point practically coincides with the maximum load. Ductility and reduction of area are small or zero. Martensitic steel is more or less fragile, of low shock resistance, very hard and unworkable. Martensitic steels are only slightly magnetic, but are capable of permanent magnetism. So long as the martensitic structure is obtained, change of carbon content has little influence on the properties.

(3) **Austenitic.** The crystals of austenite are of the tough, strong, non-magnetic  $\gamma$  iron, which usually carries in solid solution a large part and possibly all of the other chemical elements present. The strength is medium to high, 60,000 to 150,000 pounds per square inch commonly. The ratio of yield point to maximum strength is generally about 0.5, which is slightly lower than in pearlitic steels. Ductility and reduction of area are medium to large. Resistance to shock is very high. Hardness may be either low or high, according to the alloying elements and their amounts. The “necking” down of a tension test piece of austenitic steel tends to be less noticeable than with pearlitic steels, elongations being distributed over more of the length of the piece.

In the austenitic steels strength, ductility, reduction of area and hardness all increase with the carbon content, with the exception that above 0.5 per cent carbon the deformations may again decrease. Each 0.1 per cent of carbon in the austenitic structure raises both maximum strength and yield point about 6000 pounds per square inch and raises the ductility about 3.5 per cent. This statement probably holds until the carbon content is made high enough to bring out free carbide crystals of some kind. Such carbides generally lower the ductility and increase the hardness of the steel.

Mixtures of austenite-martensite occur between the purely austenitic and the martensitic steels of an alloy series. The properties are then intermediate between the properties of these two structures. Similarly, there may occur steels apparently mixtures of martensitic

and pearlitic structures, though in this case the real structure is troostitic. It is not worth while to consider the troostitic structures in alloy steels as a fourth structural type of equal importance with the pearlitic, martensitic and austenitic structures.

**383.** The alloying elements used in alloy steels may be classified according to their general methods of action as follows:

**Class 1. The main effect is depression of the critical temperatures.** The alloying element acts toward carbon as iron does. The element itself is isomorphous (forms solid solutions) with all the allotropic forms of iron, and its carbide is isomorphous with  $\text{Fe}_3\text{C}$ . The depression of the critical temperature, increasing with the amount of the alloying element, makes  $\alpha$ ,  $\beta$  and  $\gamma$  iron in succession the form of iron normal at ordinary temperatures. The steels are then successively pearlitic, martensitic and austenitic in normal structure as the amount of alloying element increases. Examples of alloying elements in this class are **manganese, nickel** and probably **cobalt**.

**Class 2. The alloying element depresses the critical temperatures and also forms a special carbide.** The action is in general the same as in class 1, but the special carbide, which is formed at high temperatures, is not to any appreciable extent isomorphous with  $\text{Fe}_3\text{C}$ , and the alloying element in forming its own carbide takes more or less carbon from solution in  $\gamma$  iron. The structures are as in class 1, save that in the austenitic steels crystals of the carbide of the alloying element are found mixed with the austenite crystals. **Chromium** is the element showing this action in commercial alloys.

**Class 3. The critical temperatures are only slightly depressed and special carbides are formed.** The alloying element is isomorphous only to a limited extent with  $\alpha$ ,  $\beta$  or  $\gamma$  iron; its carbide is not to any great amount isomorphous with  $\text{Fe}_3\text{C}$ ; the alloying element will take carbon away from iron to form its own carbide. The normal structures obtained with increasing amounts of the alloying element are at first pearlitic, then pearlitic plus the special carbide, then  $\alpha$  ferrite plus the special carbide. Examples of such elements are **tungsten, molybdenum, vanadium** and probably **aluminum, titanium** and **copper**.

**Class 4. The alloying element acts toward iron in much the same manner as does carbon.** In the preceding three classes the alloying element has been functioning somewhat like iron, or has been replacing

iron in the steel. In this class the alloying element takes the function of carbon. The critical temperatures are not much affected, but if anything are slightly raised. The alloying element may have greater chemical affinity for iron than carbon has. If so it tends to displace carbon, and if present in sufficient amount will throw carbon out as graphite. The structures are pearlitic so long as graphite is not formed, then pearlitic plus graphite or ferrite plus graphite. When graphite is formed these steels are too brittle to be of any engineering use. Examples of such elements are **silicon** and **boron**. It may be that carbon displaces the alloying element from the austenite solution. We get then in the final structure crystals of chemical compounds of iron and the alloying element. With increasing amounts of alloying element the structures are pearlitic, then pearlitic plus free crystals of the special compounds. Examples are **phosphorus** and **nitrogen**. Since the result in this case spoils the steel, there are no commercial alloy steels with phosphorus and nitrogen.

**384.** The above classification has briefly outlined the effects which the various alloying elements have when added one at a time to the carbon steels, making three component, or "ternary," steels. Frequently, however, we make four or even five component steels. In such a case the effects of the alloying elements are approximately additive. If the two or more alloying elements are taken from the same class, their effect is nearly the same as using some equivalent larger quantity of any one of them. If they are taken from different classes their effects are superposed. For example, molybdenum may replace tungsten, the weight percentage of molybdenum required for a given effect being about  $\frac{1}{4}$  of that of tungsten for the same effect. Again nickel-chromium steels are similar to nickel steels. They are more easily heat treated for high strength because in causing response to heat treatment the effects of nickel and chromium are additive. Chromium without the nickel would give too great hardness, while nickel alone would not bring out the highest strength. It will become evident that there are individual effects of the alloying elements always to be considered as well as their general effects.

**385.** The normal structures, that is, the structures in which the steels remain at the end of the usual manufacturing processes, are well shown for the ternary alloy steels by means of Guillet diagrams, so

graphist who has worked the Guillet diagrams for

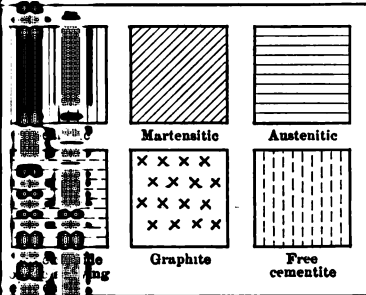
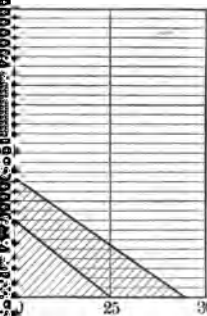


FIG. 153.

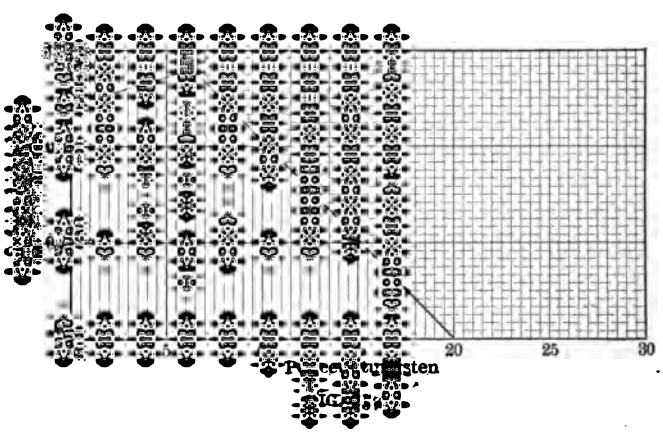
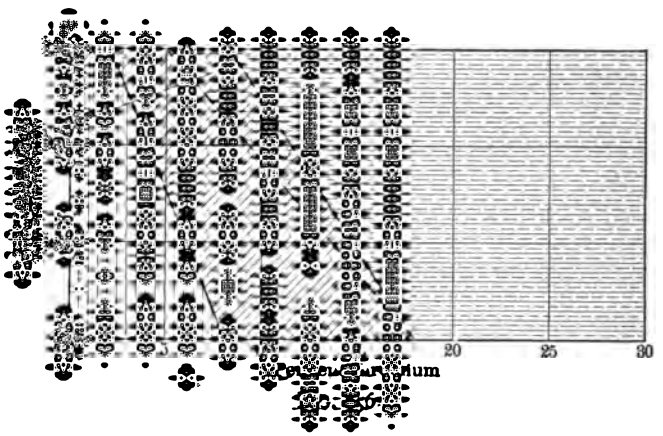
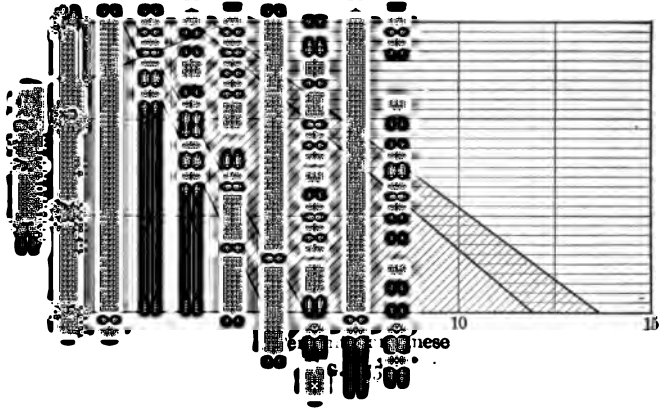
used for structure types in Guillet diagrams, Figs. 154 to 160.

at changes due to carbon values given in Section 382. 2 showing the maximum in normal structures with

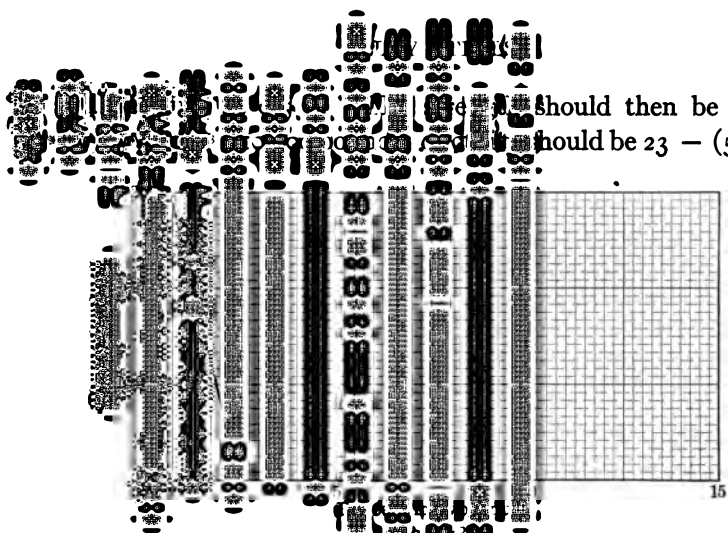


may be able to estimate alloy steel of any composition. Properties of a nickel steel with Figs. 161 and 162, for 3.5 Ni square inch and ductility 54, this steel is shown to

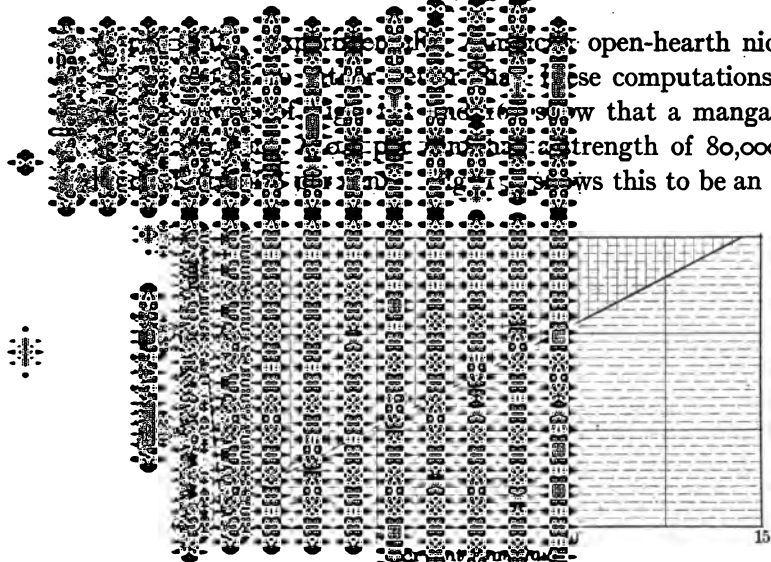
# INSTRUCTION



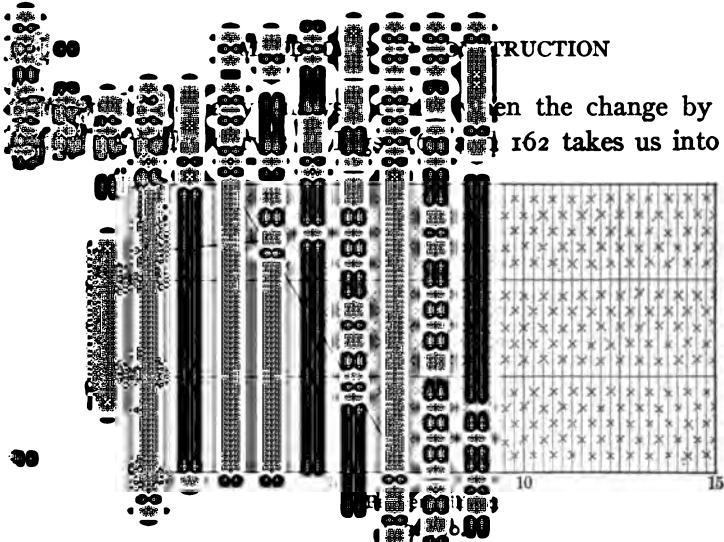
should then be  $80,000 +$   
 should be  $23 - (5 - 2) 3.5$



open-hearth nickel steels  
 these computations indicate.  
 show that a manganese steel  
 strength of 80,000 pounds,  
 shows this to be an austenitic



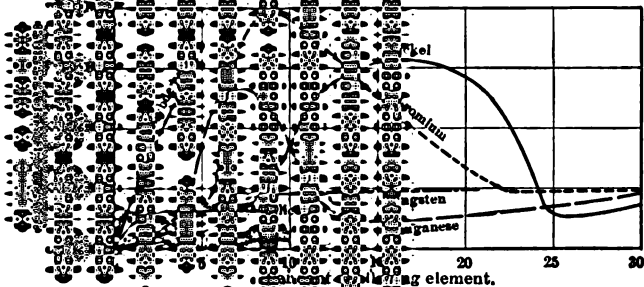
the carbon content is raised.  
 Then the strength will  
 be increased, and ductility 8 +  
 are easily obtained in the  
 properly handled in manufacture.  
 the normal properties of a



# INSTRUCTION

When the change by which we  
 162 takes us into a field of

we enter the diagrams. The  
 For instance, the steel of  
 tensitic by Fig. 155. If we  
 the new steel will be normally  
 properties of the second from



with 0.2 per cent carbon, in normal  
 make a reasonable guess at  
 Mn, 1.0 per cent C steel from  
 high carbon content. Again,  
 tungsten or molybdenum, large  
 the formation of large amounts  
 the chances for effects of carbon in



Section 382 may no longer be properly applied. If one keeps careful guard against its limitations, however, the method above outlined

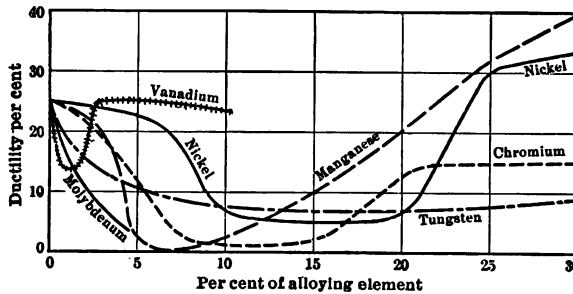


FIG. 162.

Ductilities of alloy steels with 0.2 per cent carbon, in normal condition.

for approximating to the properties of any normal ternary alloy steel will be found exceedingly useful.

**386.** The other engineering properties of the steels are intimately connected with the strength and ductility. Thus hardness is nearly proportional to maximum strength. Manganese steels are exceptional in being harder at the same strength than the comparable chromium and nickel steels. Also the martensitic steels are harder in ratio to their strengths than pearlitic or austenitic steels. Toughness and shock resistance are closely tied with ductility. A structure containing free crystals of any brittle substance may show a fair ductility under slow loading and be brittle under shock. Such an action we find in ordinary carbon steels from the presence of free crystals of  $\text{Fe}_3\text{P}$ ,  $\text{Fe}_3\text{C}$  and segregated  $\text{MnS}$ ; and it may occur in alloy steels from the free carbides of the alloying elements. Thus the high manganese and high nickel austenitic ternary steels have large shock resistance, having no free carbides in their structure, while the otherwise comparable high chromium steels are brittle from the special carbides.

**387.** Heat treatment of alloy steels is in principle the same as that of the carbon steels. In general, heat treatment is easier in the alloy steels than in the carbon steels. The reasons for this are: (1) that in many cases the critical temperatures are lower, and hence the changes of structure at the critical temperature are slower and more easily controlled, according to the general law that chemical re-

actions of the same kind are slower the lower the temperature at which they take place; (2) substances in solution in the various forms of iron, especially  $\gamma$ , seem to make slower the changes of the allotropic forms.

Any alloy steel that is normally pearlitic can be made martensitic (hardened) just as carbon steels are hardened, and rather more easily. Any alloy steel that is normally martensitic can be made austenitic by quenching from a bright red or an orange heat. Austenitic structure, as is well known, cannot be commercially obtained in the ordinary carbon steels. In the alloy steels, if sufficient amounts of the right alloying elements are present, a steel normally pearlitic can be made austenitic by quenching from an orange heat, the change from  $\gamma$  to  $\beta$  iron being so slowed by the alloy that it can be suppressed. On account of the holding of the austenitic structure by quenching, it not infrequently occurs that the alloy steels are made softer, instead of harder, by the quenching; in this case ductility will rise and strength may rise, to an astonishing extent.

**388.** With all those alloys which may form special carbides there enters into the heat treatment a peculiarity of  $\gamma$  iron. With rise of temperature the solubility of the other chemical elements in  $\gamma$  iron increases quite rapidly. Hence any alloy steel, held at a yellow heat for a sufficient time for this solubility to take effect, will reduce toward a pure austenite structure, with everything (save sulphur and oxygen) dissolved in the  $\gamma$  iron. If the subsequent cooling from the yellow heat be fairly rapid, the dissolved substances either will not precipitate out at all, or they will separate out with the formation of crystal systems differing considerably from the systems which would normally be formed with exceedingly slow cooling. On this peculiar action of the  $\gamma$  iron is based the heat treatment of the high-speed tool steels, which has as its most important part a fairly rapid cooling subsequent to a heating to a yellow or even a white heat.

The stages in the heat treatment of a high-speed tool steel may be illustrated by Figs. 163 and 164. Fig. 163 shows a high tungsten steel in its normal condition of pearlite and tungsten carbide. This steel is not a high-speed tool steel, but is given as an exaggerated case of the normal structure of such steels. The high-speed tool steels may contain 10 to 20 per cent tungsten, 1 to 5 per cent chromium and 0.5 to

amounts of molybdenum and steel is shown in the left

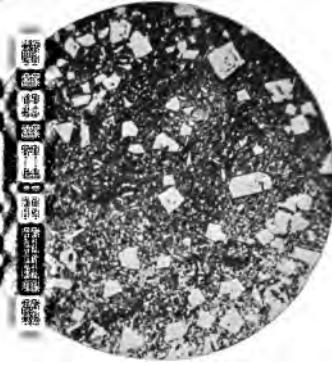
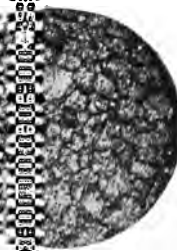


FIG. 163.

W = 39.96, C = 1.70. Normal structure (steel not hot worked); large crystals of tungsten carbide in pearlitic ground-mass. Magnified 200 diameters. From Guillet.

the high-speed steels the carbides are shown in Fig. 103 for



Left, normal structure and carbides. Right, after air quenching. Magnified 200 diameters.

On sorbite and troostite maxima the formation of carbides on sorbite is greater, so that with greater amounts of reheating the hardened structure formation continues, but

the rate of change is very much less than in the tempering of carbon tool steel. While the martensite changes toward sorbite, tending to soften the steel, some of the remaining austenite changes to martensite, restoring the hardness. This action may continue in some cases for nearly two weeks at about a dull red heat before the austenite store is exhausted and the last martensite has changed to a softer structure. At the same temperature a carbon tool steel would have become completely soft during the time of heating. The high-speed tool steels never have enough hardness to be suitable for the fine work of "finish" cuts; but the hardness they have is enough for "roughing" work and their toughness, and ability to hold their hardness when hot, enable them to take large cuts at high speeds for long times. This would be impossible with the carbon tool steels.

**389.** If a high-speed tool steel, or any similar alloy steel, is heat treated for hardening by the same process that is used for carbon tool steel, namely, water quenching from  $800^{\circ}$  to  $850^{\circ}$  C., the result will be practically the same as with the carbon tool steel. When heated between  $800^{\circ}$  and  $850^{\circ}$  C. a steel such as shown in Fig. 163 or 164 changes its pearlite areas to austenite, while the free crystals of tungsten and chromium carbides are not affected. The temperature is too low to cause their solution in the austenite. On quenching, the austenite will become martensite as in carbon steels. The carbide crystals will persist. The final structure will be martensite plus carbide crystals, hard and brittle. On reheating or tempering the martensite will decompose as it does in the carbon steels. There is, therefore, no advantage in using the high-speed tool steel if we give it the same heat treatment as carbon tool steel. The high-speed steel should always be heated, before hardening, to a high enough temperature to cause the absorption of the special carbides in the austenite. Such high temperatures as are needed are less dangerous in alloy steels than in carbon steels as causes of overheating or burning.

**390.** A large group of alloy steels are used as "structural" or "machine" steels with heat treatment to bring out the highest strength combined with ductility. The heat treatment closely parallels the similar treatment of the carbon steels. The heat treatment is usually more easily controlled and more reliable in the alloy steels because of

lower critical temperatures or of slowing up of austenite decomposition by the alloying elements. Steels with low critical temperatures give less trouble with warping and distortion during heat treatment than do steels with high critical temperatures. On the other hand, these same alloying elements may harden and strengthen the austenite so that hot working of the alloy steels is more difficult than that of carbon steels.

A very interesting set of structural and machine steels is now coming into use, with heat treatments similar to the high-speed tool steels. Their compositions are such that they are "self-hardening" by air cooling. They differ from the high-speed tool steel in that they go further through the *A-M-T-S* change during air cooling. Austenite and martensite disappear during the cooling and the structure becomes troostitic. These steels are very free from warping and other cooling stress troubles.

**391. The more common alloying elements and their uses may be listed as follows:**

**Manganese** in slight amounts, about 0.5 per cent, is used as a cleaning agent (deoxidizer) in practically all steels. With more than 1 per cent of manganese brittleness begins and from 5 to 10 per cent, the martensitic range, the manganese steels are too hard and brittle to be of any use. The steel known commercially as manganese steel carries 12 to 14 per cent manganese and about 1 per cent carbon. Its normal structure is austenitic, on the verge of becoming martensitic. It is remarkable for its combination of hardness and ductility. It is usually cast to form, though it can be hot worked at a yellow heat. It is softened by quenching from a yellow heat, becoming entirely austenitic. It is hardened by very slow cooling from high temperatures, becoming more or less martensitic. Typical uses are for railroad rails on sharp curves, frogs, switches, crossings, etc.; jaws of rock crushers; and any similar places requiring the highest resistance to simultaneous shock and abrasion.

The most common **nickel** steel has 2.5 to 3.5 per cent nickel and 0.2 to 0.3 per cent carbon. It is a structural steel replacing carbon structural steel where dead weight of structure must be kept down, as in large bridges. When so used it is not heat treated at present. Heat treatment undoubtedly would be of advantage and will be applied when

commercial methods of heating and quenching such large pieces have been worked out.

"Invar" steel, so called because its coefficient of expansion at ordinary temperatures is very small, contains 36 per cent Ni and 0.2 per cent C. It is used in making surveyor's tapes or any precision instruments which must not vary in size with temperature. Another steel with a special coefficient of expansion is "platinite," Ni 46 per cent, C 0.15 per cent, replacing platinum where wires must be sealed into glass.

**Chromium**, alone or in combination with other alloying elements, is used as a hardener. It has the disadvantage of making hot working difficult. Armor plate and armor piercing projectiles are made of nickel-chromium steel.

**Tungsten** enters into all the air-hardening steels. **Molybdenum** may be used in the same way but is more expensive. **Cobalt** is coming into use in making high-speed tool steels.

**Vanadium** has first of all a remarkable cleaning and scavenging effect upon liquid steel. This action alone, even when no vanadium remains in the finished steel, has considerable beneficial effect upon the properties. Vanadium in the steel hardens and strengthens the steel with almost as sharp an effect as that of carbon in pearlitic steels, and with far more effect per unit of alloying element added than has any other element except carbon. It has the advantage over carbon that it does not so rapidly lower the ductility. In some unexplained way vanadium also causes the structures of the steel to be unusually fine grained. The net result is that vanadium steels are remarkable for strength and toughness. The usual upper limit of vanadium remaining in the steel is 0.2 per cent, and there should be less the higher the carbon. The limit of useful addition of vanadium is 1 per cent with zero carbon. See the Guillet diagram of Fig. 159.

Commercial "vanadium" steels are usually chrome-vanadiums; and it may be that no vanadium remains in the finished steel, all having been used in the scavenging effect.

**Titanium** has a scavenging effect in the liquid steel. It is the only scavenger which will control the injurious nitrides. The titanium alloy steels seem to be unusually free from segregation in the

ingot. On these accounts titanium has proven valuable in the making of steel rails, because by stopping segregation it makes possible the use of a higher carbon content.

**Aluminum** in amounts of a few ounces to the ton of steel may be added in the ladle to stop wildness, the violent evolution of gases as the melt solidifies. As alloying material aluminum has no value; it causes ultimately both hot and cold shortness while exerting no beneficial effect.

**Copper** has the same beneficial effects as nickel for making structural steel, and should be a cheaper alloying material. That it has not come into commercial use is because of an unfounded belief that it causes hot shortness. Since this is now proven untrue we may look forward to an extensive use of copper as an alloying element, both by itself and in combination with nickel, chromium or vanadium.

**Silicon** may be used practically to replace carbon in the carbon steel so far as normal properties are concerned. Thus a steel with Si 1.0 per cent, C 0.2 per cent makes an excellent structural steel. Silicon raises the electrical resistance of steels very sharply, but does not increase the magnetic resistance. Therefore a silicon steel with Si 2.7 to 3.9 per cent and C as low as possible is used to make electrical sheet steel, on account of its low hysteresis and eddy current losses. A steel of Si 0.8 to 1.3 per cent with C 0.7 to 0.5 per cent is used for springs. Among the best of spring steels is the "silico-manganese" containing Si 2.0 per cent, Mn 0.7 per cent and C 0.5 per cent.

**392.** Below are tabulated the chemical analyses, heat treatments, properties and recommended uses of a considerable number of typical alloy steels. The data was taken generally from advertising literature of various steel companies. The properties are those of pieces heat treated in test piece size, and are comparable with those quoted for carbon steels in Section 333. The recommended uses are those given by the steel makers. We have made no recommendations of our own, because possible uses are evident from the properties. By comparisons with the lists it will be evident that some of the quenching temperatures were too low to bring out the best results possible. See Sections 299 to 301.

## NICKEL STEELS

Composition.	Heat treatment. Temperatures are °C. Temp. = tempered.	Tensile properties.				Recommended uses, and remarks.
		Strengths in 1000's lbs. per sq. in.		Deforma- tions 2 in. L X 0.5 in. D.		
		Yield	Max.	Elong.	Red. area.	
C 0.20; Mn 0.65; Si 0.25; Ni 3.5	Annealed..... Core of case-hardened bar	60 150	90 175	35 10	65 40	{ Case hardened for gears, transmission shafting.
C 0.30; Mn 0.65; Si 0.25; Ni 3.5	{ 760° oil; temp. 205° 1 hr. 760° oil; temp. 425° ½ hr. 760° oil; temp. 540° ½ hr. 760° oil; temp. 675° ½ hr.	215 135 100 70	240 150 120 95	8 16 21 25	25 50 58 70	{ Drive shafts, crank shafts, forgings. (Normal condition of steel.)
C 0.40; Mn 0.65; Si 0.25; Ni 3.5	{ 745° oil; temp. 315° 1 hr. 745° oil; temp. 540° ½ hr. 745° oil; temp. 675° ½ hr.	200 130 85	220 145 110	12 18 24	42 52 65	{ Drive shafts, crank shafts, general forgings.
C 0.20; Mn 0.45; Si 0.25; Ni 4.75	{ Case hardened..... }	110 130	130 180	8 4	8 4	{ Range of properties in case- hardened stock.
C 0.40; Mn 0.50; Si 0.25; Ni 4.75	{ Oil quenched and temp.. }	110 120	115 140	21 18	60 50	{ Range of properties possi- ble (?)
C 0.20; Mn 2.5; Si 0.20; Ni 32.5	As rolled..... Annealed.....	50 40	90 95	40 60	60 75	{ Non-magnetic, non-corrosive normally austenitic steel, of high shock resistance. Used for valves, ignition tubes, etc., in gas engines.
C 0.25; Mn 0.86; Ni 3.45	{ Normal, as rolled..... }	69	101	31	60	
C 0.14; Mn 0.63 Ni 3.64	{ 760° oil; temp. 650°..... 815° water; temp. 100°.... }	67 175	92 198	33.5 15	72.4 54.6	
C 0.35; Mn 0.45; Ni 3.39	{ 815° water; temp. 220°.... 815° water; temp. 480°.... }	291 172	307 189	10 15.5	36.3 55.5	
C 0.25; Mn 0.65; Ni 3.5	{ 845° water; temp. 100°.... 845° oil; temp. 315°..... 870° brine..... }	231 152 197	255 231 271	14 12.9 12.2	50.7 54 48	



## NICKEL-CHROMIUM STEELS

Composition.	Heat treatment. Temperatures are °C. Temp. = tempered.	Tensile properties.				Recommended uses, and remarks.
		Strengths in 1000's lbs. per sq. in.		Deforma- tions 2 in. L × 0.5 in. D.		
		Yield	Max.	Elong.	Red. area.	
C 0.30; Mn 0.50; Si 0.25; Ni 3.5; Cr 1.5	775° oil.....	265	280	10	35	Hard to forge. } Differential and transmis- sion gears. } Drive shafts, crank shafts, rods, general forgings. (Normal condition of steel.)
	775° oil; temp. 205° 1 hr.	230	240	12.5	45	
	775° oil; temp. 425° ½ hr.	165	190	13.5	55	
	775° oil; temp. 540° ½ hr.	120	135	18	65	
	775° oil; temp. 675° ½ hr.	70	105	25	68	
C 0.40; Mn 0.50; Si 0.25; Ni 3.5; Cr 1.5	760° oil; temp. 205° 1 hr.	250	260	11	40	Hard to forge. } Differential and transmis- sion gears. } General automobile forgings.
	760° oil; temp. 315° ½ hr.	225	235	11.5	45	
	760° oil; temp. 425° ½ hr.	180	205	12.5	52	
	760° oil; temp. 675° ½ hr.	75	115	24	65	
C 0.50; Mn 0.45; Si 0.25; Ni 1.75; Cr 1.0	775° oil; temp. 205° 1 hr.	235	265	8	25	Hard to forge. } Differential and transmis- sion gears. } Shafting and general auto- mobile forgings.
	775° oil; temp. 315° ½ hr.	205	225	11.5	45	
	775° oil; temp. 425° ½ hr.	160	180	13	47	
	775° oil; temp. 675° ½ hr.	70	90	30	70	
C 0.30; Mn 0.50; Si 0.25; Ni 3.0; Cr 0.60	745° oil; temp. 205° 1 hr.	220	240	12.5	48	Easy to forge. } Transmission and differen- tial gears; drive shafts. } Drive shafts; general auto- mobile forgings.
	745° oil; temp. 315° ½ hr.	185	200	13.5	56	
	745° oil; temp. 425° ½ hr.	140	150	20.5	63	
	745° oil; temp. 675° ½ hr.	95	120	22	65	
C 0.40; Mn 0.50; Si 0.25; Ni 3.0; Cr 0.60	745° oil; temp. 205° 1 hr.	250	270	12	43	
	745° oil; temp. 675° ½ hr.	85	105	25	65	
C 0.20; Mn 0.40; Si 0.25; Ni 3.0; Cr 0.60	Annealed.....	50	75	25	65	} Case hardening stock.
	Core of case-hardened bar	200	215	13	45	
C 0.20; Mn 0.65; Si 0.25; Ni 1.5; Cr 0.5	Annealed.....	45	70	25	65	} Case hardened for gears, transmission shafts.
	Core of case-hardened bar	125	150	15	40	
C 0.30; Mn 0.65; Si 0.25; Ni 1.5; Cr 0.5	790° oil; temp. 315° 1 hr.	200	205	8	35	} Transmission gear shafts. } Crank shafts, connecting rods, axle shafts, general forgings.
	790° oil; temp. 425° ½ hr.	140	175	13.5	53	
	790° oil; temp. 540° ½ hr.	105	135	17.5	60	
	790° oil; temp. 650° ½ hr.	85	100	25	65	
C 0.40; Mn 0.65; Si 0.25; Ni 1.5; Cr 0.5	775° oil; temp. 315° 1 hr.	230	250	7.5	30	} Transmission gear shaft. } Crank shafts, connecting rods, axle shafts, forgings.
	775° oil; temp. 425° ½ hr.	170	195	12.5	45	
	775° oil; temp. 540° ½ hr.	120	145	16	55	
	775° oil; temp. 675° ½ hr.	90	105	22	62	

## NICKEL-CHROMIUM STEELS (Continued)

Composition.	Heat treatment. Temperatures are °C. Temp. = tempered.	Tensile properties.				Recommended uses, and remarks.
		Strengths in 1000's lbs. per sq. in.		Deforma- tions 2 in. <i>L</i> × 0.5 in. <i>D</i> .		
		Yield	Max.	Elong.	Red. area.	
C 0.37; Ni 2.9; Cr 1.04	805° oil; temp. 650°..... 815° oil; temp. 595°.....	134 172	175 188	18.5 16	61 56	} Gears and springs.
C 0.45; Ni 2.0; Cr 1.0	} Oil temp. and annealed...	157	202	8	20	
C 0.25; Ni 2.0; Cr 1.0	} Oil temp. and annealed...	112	146	12	30	
C 0.25; Ni 3.0; Cr 1.50	} Annealed.....	128	143	15	65	
C 0.50; Ni 2.0; Cr 1.0	} As rolled.....	74	103	27	64	
C 0.30; Ni 2.0; Cr 1.0	As rolled..... Oil temp. and annealed...	78 101	119 123	18 25	45 55	
C 0.40; Mn 0.18; Ni 2.1; Cr 0.80	(As rolled..... 815° oil; temp. 620°..... 815° oil; temp. 315°..... Annealed 620°.....	134 130 237 157	170 148 264 177	12.8 17.5 10 15.5	37 40 34 48	

## NICKEL-VANADIUM STEELS

Composition.	Heat treatment. Temperatures are °C. Temp. = tempered.	Tensile properties.				Recommended uses, and remarks.
		Strengths in 1000's lbs. per sq. in.		Deforma- tions 2 in. <i>L</i> × 0.5 in. <i>D</i> .		
		Yield	Max.	Elong.	Red. area	
C 0.34; Mn 0.17; Ni 3.88; V 0.12?	{As rolled..... 815° oil; temp. 620°..... 815° oil; temp. 315°.....	65	96	27.3	54	
		83	114	16.5	51	
		132	148	15.5	55	
C 0.24; Mn 0.72; Ni 3.33; V 0.12	{As rolled..... 870° oil..... 870° water..... 870° brine; temp. 205°....	85	110	27	64	
		159	224	11.6	36	
		116	262	14.5	52	
		204	260	15.2	57	
C 0.33; Mn 0.16 Ni 3.40; V 0.24	{As rolled..... 815° oil; temp. 620°..... 815° oil; temp. 315°.....	110	148	17.8	40	
		132	139	21	61	
		184	190	15	55	

## CHROME-VANADIUM STEELS

Composition.	Heat treatment. Temperatures are °C. Temp. = tempered.	Tensile properties.				Recommended uses, and remarks.	
		Strengths in 1000's lbs. per sq. in.		Deforma- tions 2 in. <i>L</i> × 0.5 in. <i>D</i> .			
		Yield	Max.	Elong.	Red. area.		
C 0.26; Mn 0.39; Cr 0.78; V 0.17	As rolled..... 855° oil; temp. 205°.....	99 231	152 311	20 3	64 8.2		
C 0.40; Mn 0.77; Cr 1.22; V 0.19	Annealed..... 900° oil; temp. 450°.....	76 220	112 233	26 10	62 36		
C 0.38; Mn 0.73; Cr 1.19; V 0.18	{As rolled..... 905° oil; temp. 315°..... 905° oil; temp. 455°.....	92 246 143	146 323 253	22 10.8 12.9	67 47 53		} Gears and springs.
C 0.45; Mn 0.58; Cr 2.37; V 0.30	870° oil; temp. 605°..... 870° oil; temp. 220°.....	197 309	211 327	13.2 6.0	46 16	} Gears and springs.	
C 0.36; Mn 0.21 Cr 2.78; V 0.24	{As rolled..... 815° oil; temp. 620°..... 815° oil; temp. 315°.....	134 146 220	233 161 233	4.1 20 13	8.8 56 45		

## CHAPTER XIX

### NON-FERROUS METALS AND ALLOYS

**393.** The non-ferrous metals of most importance in engineering are copper, zinc, tin, aluminum and lead. These metals are all many times as expensive as cast irons and steels. They obtain commercial use because of special properties such as resistance to corrosion, ease of machining, high electrical conductivity, etc. The alloys of these metals with one another are in general more important than the pure metals.

**394.** Copper has one allotropic form, freezing at  $1084^{\circ}\text{C}$ . in a reducing atmosphere. The liquid metal is exceedingly subject to oxidation. The  $\text{Cu}_2\text{O}$  formed is soluble in the liquid copper and the metal is very easily spoiled by it. The special property of copper of great importance is its high electrical conductivity. For electrical conductors only two metals can generally be considered, namely, copper and aluminum. Weight for weight aluminum is the better, but on equal cross-sections copper is much superior.

The degree of purity desired in copper depends upon its intended use. For electrical purposes great purity is required; for mechanical purposes it is not so necessary. Lead and bismuth, even in very minute quantities, make copper unworkable. Arsenic and oxygen are also bad, but in small amounts they seem to counteract the effects of lead and bismuth, making the metal tougher. English "tough" copper always contains some arsenic. Oxides make copper weak and brittle; phosphides and sulphides make it cold short.

The strength of copper in good castings is about 25,000 pounds per square inch and ductility about 30 per cent. By cold working such as wire drawing (refining the grain and forming amorphous material on slip planes) the strength may be raised above 60,000 pounds per square inch before ductility vanishes. At the same time the yield point, which is very low and ill-defined in the cast metal, becomes definite and approaches the maximum strength. The effects of cold working

may be removed by annealing to any temperature above  $300^{\circ}$  to  $400^{\circ}$  C. The annealing allows recrystallization of amorphous material and at higher temperatures also allows crystal growth. Whether the cooling of the metal after annealing is quick or slow makes no difference. By repeated annealings alternating with periods of cold working enormous deformations of the metal are possible.

**395. Zinc** is generally known in trade as "spelter." Together with copper it is used in making the alloys called "brass." In nearly pure form, on account of its resistance to corrosion, it is used as sheets for tank linings and as a surface coating called "galvanizing" for the protection of steel wire and sheet. Pure zinc freezes at  $419^{\circ}$  C. There are three allotropic solid forms;  $\gamma$  from  $419^{\circ}$  C. to about  $330^{\circ}$  C.;  $\beta$  from about  $330^{\circ}$  to  $150^{\circ}$  C.; and  $\alpha$  below  $150^{\circ}$  C. Cast zinc has a strength of about 5000 pounds per square inch and is brittle; rolled zinc in thin sheets has a strength of about 25,000 pounds and a ductility of 20 to 25 per cent. Rolling of zinc is possible only in a narrow range of temperature just below the  $\alpha$  to  $\beta$  change. The reason will be apparent from Fig. 165, showing the dependence of strength and ductility of rolled zinc on temperature.

Steel or iron to be galvanized is first pickled in acid to remove dirt and scale. After brushing and washing it is dipped in a second acid bath, then in a heated solution of sal-ammoniac to neutralize the acid, then in molten zinc for the zinc coating. The coating on sheet metal is made smooth and even by drawing the sheets through rolls as they come from the zinc pot.

**396. Tin** resists the action of weak acids and alkalis and is therefore used in limited amounts in the form of piping for conveying such liquids. Because of resistance to atmospheric corrosion tin is used as a protective coating on iron or steel in the form of tin-plate. The alloy of tin with copper makes "bronze."

Tin freezes at  $232^{\circ}$  C. It has two allotropic solid forms, white and

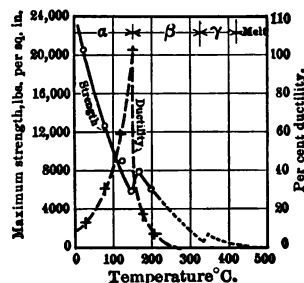


FIG. 165.

Effect of temperature on tensile properties of rolled zinc. Data of Martens, *Mitt. aus mech. tech. Versuchsanstalten in Berlin, Ergänzungsheft IV.*

gray. White tin is the ordinary form and has a strength of about 3500 pounds per square inch with moderate ductility. Gray tin does not appear to be even metallic; it occurs as a dust or powder. The transformation temperature from white to gray is  $18^{\circ}\text{C}$ . or  $64^{\circ}\text{F}$ . During a large part of the year in our climate there is a tendency for white tin to change over to gray. Ordinarily the change does not occur, because the white tin is readily supercooled, and some disturbance such as a sharp blow or contact with some of the gray form is necessary to start the change. Once started the change is rapid, especially if the temperature is about  $0^{\circ}\text{C}$ . The formation of gray tin from the white is called "tin disease"; for practical purposes the tin is spoiled.

**397. Aluminum** is used largely for electrical conductors because of its relatively high electrical conductivity and light weight. It also resists corrosion, is soft and ductile, can be easily spun, rolled, stamped and polished, and is therefore finding increasing applications in the arts in general, cooking utensils being a familiar example. It is also being used in increasing amounts in the making of alloys. It exists in one allotropic form and has a freezing point of  $657^{\circ}\text{C}$ . Its yield strength is about 6500 pounds per square inch and maximum about 15,000 pounds per square inch; ductility 25 to 30 per cent. By cold-working its yield point may be raised to about 14,000 pounds per square inch and maximum to about 28,000 pounds, ductility being 18 to 22 per cent for these values. Higher strengths may be obtained by cold working up to about 35,000 pounds per square inch, and ductility may fall as low as 8 per cent. For engineering purposes aluminum, when pure, is not especially desirable because it lacks stiffness and is so soft that machining is rather troublesome.

**398. Lead** is a heavy semi-plastic metal of very low strength (about 2000 pounds per square inch). On account of its resistance to corrosion even by strong acids it is much used for tank lining and piping in chemical works. Impure lead has a peculiar property of being workable in a semi-molten condition almost like putty. This makes possible the familiar "wiped" joint so much used in plumbing. The melting point of pure lead is  $327^{\circ}\text{C}$ . Impure lead begins to freeze near this temperature, but may not be completely frozen until the temperature is much lower. Lead is used in alloys to make brasses

free cutting, to make bearing metals, and in making fusible metals and type metal.

399. From the general consideration of strengths just given it will be seen that the pure metals themselves are of relatively little value from the standpoint of structural engineering. By mixing them to form alloys, however, great improvement of properties is possible. The most important of these alloys are those rich in copper, the brasses and bronzes or "yellow metals."

In the making of these alloys many difficulties must be overcome. The melting point of copper is  $1084^{\circ}\text{C}$ .; the melting point of zinc is  $419^{\circ}\text{C}$ . and its boiling point about  $900^{\circ}\text{C}$ . The boiling zinc burns in the air, forming clouds of zinc oxide. In making brasses, therefore, it is necessary to melt the copper first and then hold the zinc under the surface of the liquid copper while the zinc melts. Even then considerable amounts of zinc are oxidized. The zinc oxide is light and tends to float to the top of the melt, but is often trapped in the metal in appreciable quantities.

If air or oxidizing gases have access to the surface of a liquid copper-rich alloy, the copper oxidizes and the oxide dissolves in the liquid, continually exposing fresh copper for further action. Exposure to oxidation will, therefore, result ultimately in complete spoiling of the metal. A secondary reaction makes the trouble even worse. Zinc, tin or aluminum in the alloy will react with the dissolved copper oxide ( $\text{Cu}_2\text{O}$ ) to form their own oxides throughout the body of the metal. These oxides are more or less unable to escape and are even more injurious to the engineering properties of the alloy than the copper oxide itself.

400. The copper-rich alloys in the liquid condition are capable of dissolving considerable amounts of gases, just as water may dissolve air. During freezing of the alloy these gases are thrown out of solution, making the solid casting internally spongy. As a spongy casting is almost certainly badly oxidized, both its appearance and its action are well described by the word "rotten." The bubbles of gas so held in a solid casting are called "occluded" gases.

401. The remedies for oxidation and occlusion are partly preventive and partly curative. Methods of prevention include forcing the melt, preheating the metals, and heating rapidly to shorten the

time of exposure of liquid metal in the furnaces; and using fluxes such as powdered charcoal or graphite covering the surface of the metal and excluding gases. Methods of cure are: (1) "Poling," stirring with a stick of green wood. This mixes the metal, allows oxides to float up, and also deoxidizes by the charring of the wood and reaction of the oxides with the charcoal. (2) Zinc oxide to a considerable extent floats to the top of the melt and may be skimmed off. Therefore, zinc can be used as a deoxidizer, though not an extremely effective one. (3) Magnesium and aluminum are good deoxidizers, but their oxides rise so slowly in the melt that there is danger of these oxides passing over into the castings. (4) Manganese used in amounts up to 2 per cent. If in excess of the amount needed for deoxidation, manganese strengthens and toughens copper alloys by going into solid solution in the copper, so that the amount of manganese used need not be closely fixed. Manganese is added in the form of a copper-manganese alloy which is cheaper and has a lower melting point than pure manganese. (5) Silicon added in the form of silico-copper is used in quantities just sufficient to deoxidize. Any excess sharply reduces ductility, making the metal hard and brittle. (6) Phosphorus in the form of a copper-phosphorus alloy, which is largely  $\text{Cu}_3\text{P}$ , may be used in quantities a few tenths of a per cent greater than required for deoxidation. Care must be used to prevent volatilization and direct oxidation of phosphorus as with zinc. Phosphorus pentoxide ( $\text{P}_2\text{O}_5$ ) floats out of the metal. Excess phosphorus if small in amount seems to form a solid solution with copper, strengthening and toughening the metal. Larger amounts of excess phosphorus bring out a brittle eutectic structure of copper and  $\text{Cu}_3\text{P}$  which spoils the metal. (7) Fluxes may be used as deoxidizers. Charcoal or powdered graphite acts weakly in this way. Boron suboxide is a strongly deoxidizing flux and promises to have wide commercial use.

In many cases the deoxidizer used gives a name to the resulting alloy. Examples are silicon-copper, silicon-bronze, phosphor-bronze, manganese-bronze, etc. Oftentimes none of the element giving its name to the alloy remains in the finished metal.

**402.** In casting these metals and their alloys, other difficulties are encountered. With careless pouring, allowing the liquid metal to splash, bubbles of air are caught in the metal and carried into the



mold. Splashing can be largely prevented by the use of a centrifugal sprue. Instead of pouring the metal directly into the sprue it is poured into a depression in the sand, and from this depression discharges tangentially into the top of the sprue. The metal takes a helical path down the sprue, centrifugal force causing it to flow smoothly without splashing.

**403.** If the pouring temperature is too low the metal is viscous and the molds will not fill. If the pouring temperature is too high oxidation troubles in the furnaces are increased; the cooling of the metal in the molds is slow during the freezing period and the casting is coarse-grained. A pouring temperature between  $100^{\circ}$  and  $200^{\circ}$  C. above the melting point of the alloy is probably best, but the proper temperature depends upon the individual metal and casting.

**404.** If a casting freezes fairly slowly, so that the central parts are liquid an appreciable length of time after the outside is solid, segregation may occur. Especially oxides and gas bubbles will collect in certain spots. Selective freezing of solid solutions may cause the outside of a casting to run low in zinc, tin or aluminum as compared with the center. When this action is excessive it is called "liquation."

**405.** The shrinkage of the copper-rich alloys in casting is usually a little less than  $\frac{2}{10}$  inch per foot. The same general troubles with shrinkage stresses occur here as in cast irons. See sections 337, 338, 339 and 364.

**406.** The reasons for variations of properties with composition and heat treatment in the copper-rich alloys can best be studied by reference to the equilibrium diagrams, Figs. 166, 168 and 170. In a general way the diagrams for copper-zinc, copper-tin and copper-aluminum alloys are strikingly similar. In each case the first additions of alloying element to the copper go into solid solution in the copper, making crystals similar to those of pure copper in structure, but of strength and ductility increasing as the amount of dissolved substance increases. These solid solutions are called the  $\alpha$  crystals of each alloy series. The second crystal type in each alloy series is the  $\beta$  solid solution. The  $\beta$  crystals show a slightly reddish tinge on a cut surface, in contrast to the pure yellow of the  $\alpha$ . This must be qualified by remembering that pure copper is distinctly red. The  $\beta$  crystals are of high strength and of low ductility when cold, but not brittle. The  $\beta$  crystals do not

normally come down to room temperature, but decompose at a dull red heat into a mixture of  $\alpha$  crystals with  $\gamma$  or  $\delta$  forms. The  $\gamma$  crystals of each series are white in color, brittle and weak. This brittleness and weakness runs through with further increases of alloying element until we approach pure zinc, tin or aluminum. The commercial copper-rich alloys of these series are limited to the  $\alpha$  or  $\beta$  crystalline regions, the metal becoming valueless as soon as neither  $\alpha$  nor  $\beta$  crystals are present in the structure.

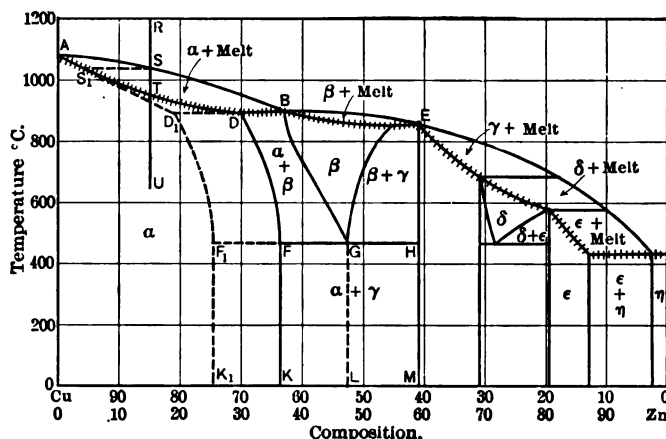


FIG. 166.

Equilibrium diagram of copper and zinc; after Shepherd, *Jour. Phys. Chem.* VIII, 1904, p. 421, modified for results of Carpenter, *Jour. Inst. Metals*, 1911, I, and 1912, I and II.

**407.** While the general nature of the copper-rich alloys can be outlined as above, the actual effects on properties of each alloying element are specific. Most important commercially of the copper-rich alloys are those of copper and zinc, generally called brasses. The equilibrium diagram is shown in Fig. 166 and the properties in Fig. 167. By comparing the two figures it will be seen that both strength and ductility rise from pure copper to the other limit of the  $\alpha$  solid solution at about 35 per cent zinc. Between 35 and 45 per cent zinc, approximately, strength rises and ductility falls. Either  $\beta$  crystals, or the eutectic of  $\alpha + \gamma$  from the decomposition of the  $\beta$  (at point G, Fig. 166), are much stronger than  $\alpha$ , but of lower ductility. Beyond 45 per cent zinc both strength and ductility decrease because of the

presence of free  $\gamma$  crystals. Beyond 60 per cent zinc the alloys are entirely worthless, having neither strength nor ductility.

**408.** Commercial brasses very generally contain lead in amounts from 0.5 to 3.0 per cent. The advantage from the addition of lead is that the metal cuts more cleanly and freely in machining. In the cold brasses the lead is not alloyed but is present as minute lead crystals scattered through the metal. The disadvantage of the lead is that it

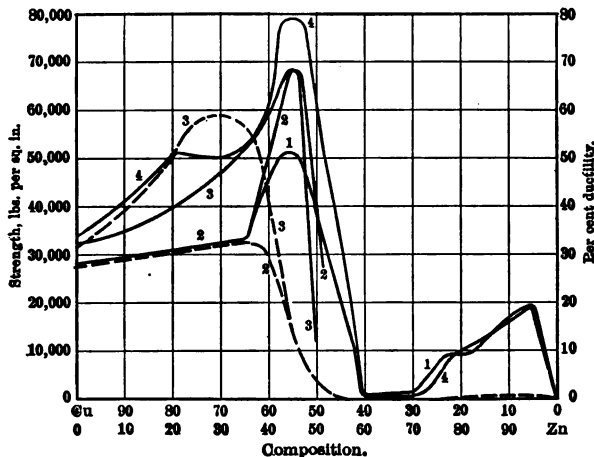


FIG. 167.

Tensile properties of copper-zinc alloys. 1, Thurston, castings; 2, Lohr, castings in graphite molds, taken out and water quenched from 700° C.; 3, Charpy, annealed brass; 4, Alloys Research Committee, worked bars. Full lines are for maximum strengths, dash lines for ductilities.

melts somewhere between 300° and 350° C. when the brass is heated, making the brass hot short. Up to the edge of the  $\alpha$  field, or about 35 per cent zinc, the brasses have enough ductility to stand cold-rolling, and hot shortness from the presence of lead is not serious. In the absence of lead these same brasses can be hot rolled as well as cold rolled. "Muntz" metals, brasses containing about 40 per cent zinc, can be hot worked even with lead in them. The hot  $\beta$  crystals have considerable ductility, and also have the power to take lead into solid solution so that it ceases to be a cause of hot shortness.

**409.** The equilibrium diagram of the copper-tin alloys, or bronzes, is shown in Fig. 168. The properties in the cast condition are given in Fig. 169. The strengths of the ductile bronzes can be increased

by cold working just as can those of the brasses. The commercial bronzes commonly carry 8 to 15 per cent of tin, approximating the boundary of the  $\alpha$  solid solution. The remarkably sonorous, hard,

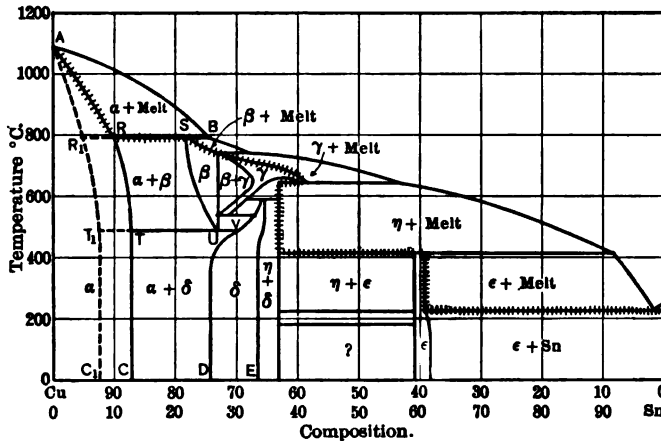


FIG. 168.

Equilibrium diagram of copper and tin; modified from Shepherd and Blough, *Jour. Phys. Chem.* X, 1906, p. 630.

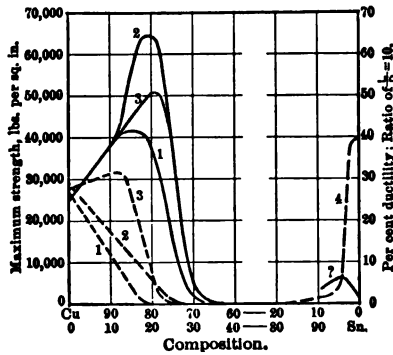


FIG. 169.

Properties of copper-tin alloys. Full lines, maximum tension strengths; dash lines, ductilities. 1, as cast in graphite molds; 2, cast and heated to red ( $600^{\circ}$ – $700^{\circ}$  C.) and water quenched; 3, cast, annealed at  $540^{\circ}$  C. one week, water quenched; 4, sand cast. 1, 2 and 3, Upton, *Jour. Phys. Chem.* IX, 1905; 4, Thurston.

brittle, moderately strong alloy known as "bell metal" contains about 22 per cent of tin and consists of a mixture of  $\alpha$  and  $\delta$  crystals. "Speculum" metal is pure  $\delta$ . Comparison of Figs. 168 and 169 shows that

both strength and ductility vanish in the absence of  $\alpha$  or  $\beta$  crystal types;  $\gamma$ ,  $\delta$ ,  $\epsilon$ , or mixtures of them, are both weak and brittle.

410. The equilibrium diagram of copper and aluminum is shown in Fig. 170 and the properties in Fig. 171. The aluminum-bronzes, as these alloys are commonly called, are white, brittle and worthless materials if the aluminum exceeds 14 or 15 per cent. Free  $\gamma$  crystals occur beyond the line  $IM$ , Fig. 170, and as soon as they occur the

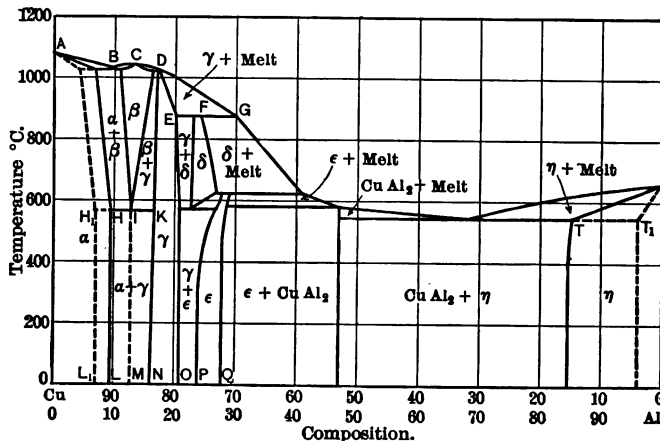


FIG. 170.

Equilibrium diagram of copper and aluminum; modified from Curry, *Jour. Phys. Chem.* XI, 1907.

metal is worthless. The strengths shown in Fig. 171 are probably considerably below those possible in completely deoxidized metal. There is reason to believe that if the liquid copper should be deoxidized with boron suboxide flux before the addition of aluminum, the 10 per cent aluminum alloy might show in castings strengths as high as 100,000 pounds per square inch. This metal is comparable in properties with good forgings of medium-carbon steel, but is of course far more expensive. It has great resistance to corrosion and is non-magnetic. These properties may make the aluminum-bronze replace steel under certain trying conditions.

411. Commercial brasses and bronzes usually contain more than one element added to the copper. To the brasses, which are principally copper and zinc, lead is added to make machining easy, tin to reduce corrosion, manganese for deoxidation and strengthening and

## INSTRUCTION

, principally copper and tin, lead for ease of machining, manganese for deoxidation and iron for strengthening. The these complex metals are fairly and by the properties of the in the proper percentages, as 166 to 171 and their discussion. 1289 and 290 is a list of some important copper-rich alloys. In symbols are used as follows:  $\sigma$  for yield point strength,  $M$  for tension strength,  $D$  for ductility, and  $A$  for

Fig. 167, 169 and 171, showing the copper-zinc, copper-tin and copper- it will be noticed that the strength of these alloys is markedly lower than that of the wrought or annealed metals. The reason for this is the failure to reach the maximum strength during cooling of the castings. The actual course of the strength of the alloy. The actual course of the strength of the alloy. Fig. 166. Suppose a brass of 60 per cent zinc, is to be cast. The actual course of the strength of the alloy. At the point  $S$  crystals of the alloy begin to form from the liquid. The composition of the first crystals is found by drawing a line of constant composition from  $S$  to the solidus line of the diagram. The first  $\alpha$  crystals contain only 5 per cent zinc. On further cooling successive layers of crystals rapidly increasing their zinc content are formed. The zinc content of successive layers increases in zinc as the temperature falls. The composition of the last layer is  $AS_1TD$  (the solidus line). As the temperature falls the alloy becomes richer and richer in zinc as the temperature falls. The composition of the last layer of the crystals must

Composition.	Treatment.	Name, properties, uses or remarks.
Cu 85-60, Zn 15-40, Pb up to 2.	Cold rolling	General limits of cold rolling brasses; cannot be hot rolled if they contain Pb.
Cu 62, Sn 1+, Zn 37.	Cold rolled	Naval brass. $Y=30,000$ , $M=60,000$ .
Cu 60-70, Zn 40-30, Pb 0.5.	Cold rolled	Sheet brass; seamless steam and water pipe.
Cu 70, Zn 29, Sn 1.	Cold drawn	Admiralty metal. Condenser tubes in salt water.
Cu 67-73, Zn 33-27, Pb traces, Fe traces.	Cold drawn	Condenser tubes in fresh water. $Y=24,000$ , $M=55,000$ , $D=32$ and $A=35$ .
Cu 62.5, Zn 35, Pb 2.5.	Cold rolled	Rod brass.
Cu 60-55, Zn 40-45.	Hot rolling	General limits for hot rolling brasses. Pb can be carried because it goes into solid solution in $\beta$ crystals.
Cu 60, Zn 40.	Hot rolled	Muntz metal. Bolts and nuts. $M=35,000$ .
Cu 60-70, Zn 40-30, Sn up to 2, Pb up to 2.	Casting	General limits for casting brass.
Cu 62, Sn 1+, Zn 37.	As cast	Naval brass. Castings where high strength is not required.
Cu 85, Zn 15.	As cast	Brazing metal; all fittings held by brazing.
Cu 88, Zn 10-8, Sn 2, Pb up to 2.	As cast	Navy gun bronze. Used for valves and fittings of great strength.
Cu 77-80, Sn 4, Pb 3, Zn 13-19.	As cast	Navy brass. For screw pipe fittings.
Cu 88, Sn 10, Mn 2.	As cast	Manganese bronze (one kind). $M=70,000$ , $D=30$ . Used for propellers.

Composition.	Treatment.	Name, properties, uses or remarks.
Cu 88.5, Zn 1.6, Sn 8.7, Fe 0.7, Pb 0.3, P trace, Mn zero.	As cast	Parson's manganese bronze. Propeller castings.
Cu 56, Zn 41, Sn 1—, Al 0.5, Fe 1.3, Mn 0.1.	As cast	Manganese bronze (ordinary). $Y=30,000$ , $M=60,000$ . Castings of great strength.
Cu 59, Zn 39, Sn 2+, Pb 0.3, Fe 0.1, P less than 0.25.	Cast, or hot rolled or cold rolled.	Tobin bronze. Cast, $M=66,000$ ; hot rolled, $M=79,000$ ; cold rolled, $M=104,000$ . Used for castings of great strength; hot rolled into rod, bolts and nuts. Non-corrosive.
Cu 85-88, Sn 7-15, Zn 0-5, Pb up to 1, P less than 0.25.	As cast	Valve bronzes for steam valves.
Cu 87-92, Sn 8-10, Zn up to 2.5, P less than 0.25	As cast, or cold rolled	Phosphor-bronze. Used for pump rods, valve springs. Resists salt water corrosion.
Cu 83, Sn 13.5, Zn 3.5.	As cast	Navy bearing bronze.
Cu } Zn } Sn } Fe 0.1 to 2.0.	As in other brasses and bronzes.	Delta, or sterro metals. Same as ordinary brasses or bronzes but stronger, because made with a zinc containing considerable iron.
Cu 97, Si 3.	As cast	Silicon-bronze. $M=55,000$ , $D=55$ .
Cu 95, Si 5.	As cast	Silicon-bronze. $M=75,000$ , $D=8$ .



vary in composition from  $S_1$  toward  $D$  in such proportions that the average composition is that of  $T$  (15 per cent). There is thus produced what is known as a "cored" crystal. Each single crystal varies in composition from center to outside to a very considerable extent.

If the freezing were slow enough for equilibrium to be maintained there would be diffusion of zinc through the solid solution so that the cores of the crystals would keep up during freezing to the same zinc content as the outsides of the crystals. In this case freezing would be complete at  $T$  to a homogeneous set of solid solution crystals each containing 15 per cent zinc throughout. Since it has been found that diffusion requires about 3 hours of time near the melting temperature for equilibrium to be obtained through large cored solid solution crystals of copper alloys, it follows that equilibrium never is attained in the making of commercial castings.

**414.** Equilibrium can be obtained by an annealing as just indicated, or by a very different procedure. If a casting is chilled, or very rapidly cooled during freezing, it freezes to a very large number of small crystals instead of a small number of large crystals. The coring is of the same kind, but equilibrium is much more quickly obtained in annealing. The engineering properties of the fine-grained structure are likely to be better than those of the coarse-grained.

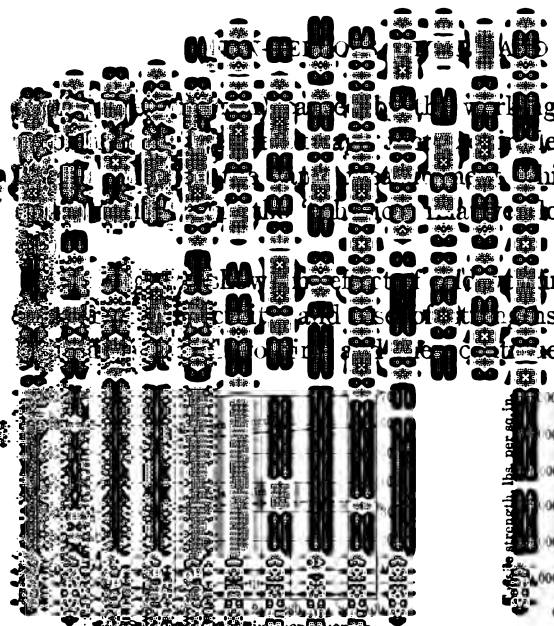
**415.** In case any portion of the melt at the end of freezing reaches the temperature level  $DB$ , Fig. 166, that portion of the melt will form  $\beta$  crystals. In cases of excessive coring of the  $\alpha$  it is very probable that this formation of  $\beta$  will occur. The effect is virtually to move the boundary line  $DFK$  of the equilibrium diagram between the fields of  $\alpha$  and  $\alpha + \beta$  or  $\alpha + \gamma$ , to the general position  $D_1F_1K_1$ . In the presence of cored  $\alpha$  crystals we are very likely to find  $\beta$  or  $\gamma$  crystals where they would not be expected from the equilibrium diagram. The line  $D_1F_1K_1$  in Fig. 166 is placed approximately as it should be for the sand casting of brasses. Similarly, on account of the same coring action the line  $RTC$  in Fig. 168, for copper-tin alloys, is displaced in commercial work to  $R_1T_1C_1$ ; and  $HL$  in Fig. 170 for copper-aluminum alloys is displaced to  $H_1L_1$ .

**416.** Cored crystals are less ductile than homogeneous crystals of the same composition. Also  $\beta$  and  $\gamma$  crystals of the copper-rich alloys are much less ductile than  $\alpha$ . Hence it follows that castings of the

copper-rich alloys are considerably less ductile than the annealed or worked and annealed metals. This shows very sharply in Fig. 169 for copper-tin properties. The dashed curve 1 is for chilled castings. Curve 2 shows that heating to red heat, then quenching, markedly raises the ductility of the castings. The effects of the annealing and quenching are: (1) diffusion reduces the coring of the  $\alpha$  crystals making them more homogeneous; (2) excess  $\beta$  tends to disappear by diffusion of tin toward the cores of the  $\alpha$  crystals; (3) any  $\beta$  remaining after the diffusion is caught by the quenching and prevented from changing to the more brittle  $\delta$  crystals as it would do in slow cooling. The castings cooled slowly enough so that they contained  $\alpha$  (cored) +  $\delta$ . All of these various changes tend toward increase of ductility so that heating to red and quenching is very beneficial to these bronzes. The full-lined curve 2, Fig. 169, shows that the strength is also improved if the composition is such that  $\beta$  crystals can be brought down by the quenching.

Curve 3, Fig. 169, is put in to show the effect of annealing to complete equilibrium. The annealing time was one week, probably much longer than was necessary. After the annealing the pieces were water quenched. The dashed curve 3 shows that when equilibrium is attained, or coring and excess  $\beta$  have disappeared, ductility rises through the  $\alpha$  solution field of copper-tin with increase of tin. This is the normal property which is entirely masked in castings by the lack of equilibrium. Similar effects from annealing are shown in copper-zinc and copper-aluminum alloys, but the ductility changes are not relatively so large as they are in copper-tin.

**417.** One of the most important methods commercially used for hardening and strengthening the copper-rich alloys is that of cold working. Cold working processes include spinning, pressing, wire drawing and rolling of rods or sheets. During the cold working the strength and hardness of the metal are greatly increased at the expense of the ductility. If the working is carried too far the metal will break. Therefore, if the amount of deformation is very large the metal must be annealed to restore ductility after which it may again be cold worked. The effect of cold working and annealing upon the properties is illustrated in Figs. 172 to 175. The amount or degree of cold working is measured by the change of length or cross-



length over the original length which has previously been elongation." (See Section

ing on electrolytic copper. The specific measure of elongation is very rapid at the beginning and then slowly through a great

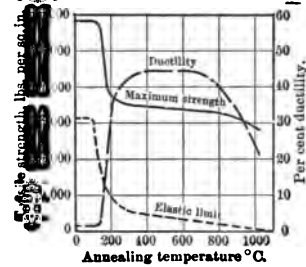
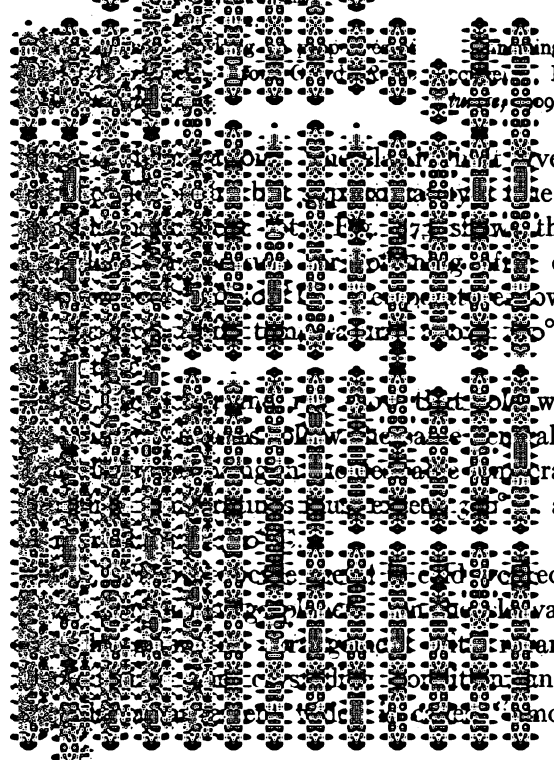


FIG. 173.

ing of cold-worked electrolytic copper. From Gard, *Revue de Metal-*



even in Figs. 172 to 175 is the elastic limit of first per cent that for pure copper the cold working should be lower than 200° C. will not 200° C. bring about no real

working and annealing of copper. The laws as for copper, the temperature of annealing. The and may with advantage

its crystals are broken along the slip planes the molecules are forced into the amorphous." The arrange-

# CONSTRUCTION

amorphous, but the liquid dif-  
 the liquid the molecules are  
 each other, a motion which is  
 solid. The amorphous material  
 by cold working is often com-  
 such comparison is improper.  
 amorphous  $\alpha$  iron molecules,  
 does not contain molecules

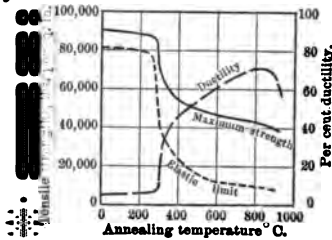


FIG. 175.

Annealing of cold-worked 67-33 brass.  
 from Gard, *Revue de Metallurgie*, 1909.

the amorphous solid formed  
 same molecular kind as the  
 same molecular kind as the  
 hardening by cold working is  
 an appreciable amount of the  
 ordinary metals is considerably  
 the same molecular type. A  
 strengthening by cold working is  
 This refining of crystal or  
 copper-rich alloys. The rapid  
 cold working is due to forma-  
 with a slight amount of amor-  
 the slowing up of the change of  
 the first stages is explained by  
 but instead increasing the  
 planes already present.  
 by annealing the first stage  
 of the amorphous material.  
 place is below 200° C. in copper,

about 300° C. in brasses and bronzes and toward 400° C. in steels. This recrystallization of amorphous material causes a very sharp change in properties as shown in Figs. 173 and 175. The second stage of the annealing, which does not occur unless the temperatures are higher than those for the first stage, is a period of crystal growth. If cored crystals were present in the metal before cold working and were broken down during cold working, the annealing also gives opportunity for diffusion toward equilibrium. The difference in form of the ductility curves in Figs. 173 and 175 is probably explained by this diffusion toward equilibrium of an originally cored crystal structure in the metal of Fig. 175.

If annealing is carried out at a temperature too near the melting point the metal may be burnt. This is especially liable to happen to brasses, in which the zinc easily oxidizes. This danger limits annealing temperatures for brasses to 700° to 750° C. for heavy pieces, and 600° to 650° C. for thin sections. The burning causes loss of both ductility and strength. This is shown at the high temperature ends of the curves in Figs. 173 and 175.

**422.** It sometimes becomes necessary to keep metal almost continually at a fairly high temperature. Such a case occurs in boilers and steam pipes. We are then interested in knowing how the properties of the metal are affected by temperature. Two kinds of action are to be expected: (1) the properties are all functions of temperature in any case even though the structure is not changed; (2) changes of structure may occur. Crystal growth without change of crystal type will go on with time if the temperatures are kept continually above 300° to 400° C. and will cause slow alteration of properties. Allotropic changes may also occur.

The variation of properties of carbon steels with temperature are indicated in Figs. 83 and 84. These figures do not indicate the deterioration that would occur by long exposures above 400° C. and the consequent crystal growth. The effect of temperature on properties of brasses and bronzes, again with the time of exposure to high temperature not long enough for marked crystal growth, is shown in Figs. 176 to 178. In these metals crystal growth begins to appear around 300° C. or 600° F. The sudden change of properties, at about 250° C. (500° F.) in navy brass, Fig. 177, is probably due to a melting

## CONSTRUCTION

to be decidedly hot short. In navy bronze. It is evident from the figures that the properties of these bronzes are questionable for uses higher than  $250^{\circ}\text{C}$ . or  $500^{\circ}\text{F}$ . In the absence of lead bronzes and brasses which hold their strength fairly well up to  $800^{\circ}\text{F}$ . or  $400^{\circ}\text{C}$ .

In machinery where lightness of weight is of the highest importance, these various "light alloys" are used. In compression, tension, and torsion, wherever moment of inertia of the member is of as much importance as the strength of the material in determining the length of the member. So far as the tensile strength of steel per unit weight is of any other metal that there is no such use of light alloys is the case where light alloys are commercially available.

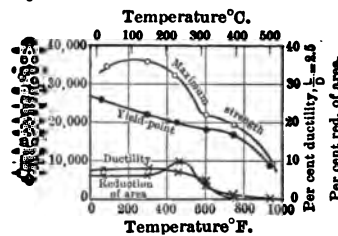


FIG. 178.

Effect of heat on properties of cast navy bronze, Cu 88, Sn 10, Pb 2. From *Valve World*, Jan. 1913.

development and application. Their use is limited. With this may be alloyed with aluminum, copper, zinc,

and others. The symbols are the same as in general early ones;

the more recent and probably better light alloys have not had their formulas published.

Composition.	Remarks.
Al 95.5, Cu 3.0, Mn 1.0, Mg 0.5.	Trade name is "Duralumin." Said to be stronger than cast iron in castings, ductile and capable when cold worked of strengths comparable with steels.
Al 92, Cu 8, Ni trace.	Generally used in automobile work. Cast, $M=18,000$ ; rolled, $M=50,000$ ; specific gravity, 2.84.
Al 90-94, Mg 10-6.	Trade name "Magnalium." Cast, $M=\text{about } 20,000$ ; $D=4$ ; rolled, $M=50,000$ , $D=3.7$ ; rolled and annealed, $M=40,000$ , $D=18$ .
Al 90-80, Zn 10-20.	Has been used in automobile work for castings. Generally discarded because of unreliability. If the zinc is high the metal spontaneously disintegrates in time. Cast, $M=23,000$ ; specific gravity = 3.1.
Al 83, Cu 2, Zn 15.	Used like the aluminum-zinc alloy above and with the same defects.

**424.** A very important group of non-ferrous alloys are those used for making bearings. We desire in a bearing the lowest possible coefficient of friction with the shaft. While lubrication continues the coefficient of friction depends much more upon the lubricant than upon the bearing metal. The bearing metal contributes to the result, however, by the shape, character and condition of its surface. The spreading of lubricant over a metal surface depends upon the surface tension of the lubricant in contact with that metal. In general the metals of high atomic weight and high density spread oil better than do metals of low density. This action has to do with the point at which lubrication fails as speed decreases or pressure increases.

Bearing metals can be divided approximately into two classes, soft and hard. The soft bearing metals will not withstand high pressure or severe service, but they may be cast into place in the bearings, cheaply installed, and their semi-plasticity enables them to shape themselves to a good fit with the shaft without any particular amount of machining. The main part of the soft bearing metals is lead or tin. Antimony (or copper) is used as a hardener. The antimony combines

with some of the tin to make a set of hard crystals which have low coefficient of friction. In the bearing metal these hard crystals are embedded in a background mass of soft metal consisting of the remaining lead and tin. In operation the wear of the bearing soon causes the shaft to run on the hard crystals of the bearing metal while any projecting points or ridges are pushed down into the ground mass so that a good fit is obtained.

The hard bearing metals, used for high pressures and severe service, are generally copper alloys containing also lead, tin and zinc. These copper alloys do not contain the mixture of hard and soft crystals and are too strong and stiff to fit themselves to place. They must be machined and fitted.

For pressures so high that the soft metals might squeeze out, bearings are sometimes made of a strengthening cage of brass or bronze in the form of a grid, and the soft bearing metal is cast into the spaces of the grid.

Following is a list of compositions of a few typical bearing metals.

Composition.						Name and remarks.
Cu	Sn	Pb	Zn	Sb	P	
3.5	55	23.5	....	18	....	"Babbitt" metal.
....	10	70	....	20	....	} Lining metals used by the Eastern R.R. of France.
....	12	80	....	8	....	
6	83	....	....	11	....	
10	....	65	....	25	....	} Navy bearing bronze.
83	13.5	....	3.5	....	....	
80	10	9.5	....	....	0.7 to 1.0	} Phosphor bronze. } Pennsylvania R.R. "K" bronze } "B" bronze was found to be best. "B" bronze. } "Plastic" bronze. } "Plastic" bronze developed from Pennsylvania R.R. experience.
77	10.5	12.5	....	....	....	
77	8	15	....	....	....	
80	....	20	....	....	....	
65	5	30	....	....	....	

425. There are a number of miscellaneous non-ferrous alloys of commercial importance, some of which are listed on page 299 with their special properties or uses.



Composition.	Name and remarks.
Cu 26.5, Ni 72.0, Fe 1.5.	"Monel" metal is a natural alloy of copper and nickel. It has strength like that of machine steel, is slightly magnetic and is non-corrosive. In castings, $M=70,000$ to $80,000$ ; rolled $M=100,000+$ .
Cu 82, Mn 15, Ni 2.3, Fe 0.6.	"Manganin" is used as an electrical resistance wire of high resistance and very low temperature coefficient of resistance.
Cu 60, Ni 40.	"Constantin" has properties like Manganin. Used also for thermo-couples.
Ni 18-25, Zn 20-30, Cu remainder. Sn may replace Zn if Ni is low.	"German silver" is used for resistance wire and for many small articles in arts on account of bright color and resistance to corrosion.
Pb 67, } Sn 33, } Pb 50, } Sn 50, } Pb 33, } Sn 67, }	Soft Medium Hard } Solders. Hard solder is the strongest and most expensive.
Pb 83-80, } Sb 17-20. }	Type metal.
Pb, Sn, Bi, Cd.	Mixtures of these four elements in more or less equal amounts make fusible metals used in fire alarm systems, automatic sprinklers, safety plugs in boilers, etc. By changing the composition the melting point can be altered to any desired value within wide ranges.

## CHAPTER XX

### CEMENT AND CEMENT TESTING

**426.** Cements are materials used for binding together stone or brick in structures. There are two main classes of cements: (1) lime cements, which will not "set" under water and if hardened in air and put under water will disintegrate; and (2) cements which will harden and stay hard under water and are therefore called hydraulic cements. Hydraulic cements harden equally well in air or water.

**427.** Lime cements undergo a large volume change in hardening. In order to mask this volume change large amounts of sand must be used with lime in making mortar. Lime is in any case much weaker than the hydraulic cements, and the addition of large amounts of sand in the lime mortars brings about still further weakening. Because of these properties lime is now used chiefly in light work or interior work such as plastering. Mixtures of lime and hydraulic cements (Portland) are used in light outdoor construction, giving greater strength and resistance to weathering than lime mortars, and being cheaper and working better with a trowel than hydraulic cement mortars.

Lime itself is  $\text{CaO}$  or "quicklime." It is made by burning limestone rock,  $\text{CaCO}_3$ , to drive off  $\text{CO}_2$ . Either in manufacture or just before use the quicklime is changed to "slaked lime" or lime hydrate,  $\text{Ca(OH)}_2$ , by reaction with water. The wet mortar is a paste of lime hydrate. Hardening is mainly due to the crystallization of the lime hydrate. Superficially, by reaction with the  $\text{CO}_2$  of the air, some  $\text{CaCO}_3$  is also formed.

**428.** By far the most important hydraulic cement is the artificial material known as "Portland" cement. This material has now so crowded out competing materials from the market that to the engineer the unqualified word cement means Portland cement. The constituents of Portland cement are  $\text{CaO}$  (lime),  $\text{SiO}_2$  (silica),  $\text{Al}_2\text{O}_3$  (alumina) and  $\text{Fe}_2\text{O}_3$ . The proportions in which these constituents occur are approximately  $\text{CaO}$  65,  $\text{SiO}_2$  25 and  $\text{Al}_2\text{O}_3 + \text{Fe}_2\text{O}_3$  10 per cent by

weight. MgO may replace CaO to a limited extent, approximately 5 per cent. In this limited amount MgO seems to form chemical compounds similar to those made by CaO and isomorphous with them.  $\text{Al}_2\text{O}_3$  and  $\text{Fe}_2\text{O}_3$  do not act in an entirely similar manner but are enough alike so that they are generally considered to have the same functions.

**429.** Rocks containing these constituents of Portland cement are ground, usually dry, to flour-like fineness, mixed in proper proportions and "burnt" in slanting rotary kilns. The temperature in the kilns is not high enough to melt the material of the cement, although a portion of it melts and so helps the rest to aggregate into "clinker." In the kilns, after driving out moisture and  $\text{CO}_2$  from the ground rock, chemical reactions occur between the CaO and the  $\text{SiO}_2$ ,  $\text{Al}_2\text{O}_3$ , etc. These reactions depend upon the time and temperature of burning, and the fineness of grinding and thoroughness of mixing before burning. The reactions are essential to the making of Portland cement. The chemical compounds formed are typically tri-calcic silicate ( $3 \text{ CaO}$ ,  $\text{SiO}_2$ ), di-calcic silicate or calcium orthosilicate ( $2 \text{ CaO}$ ,  $\text{SiO}_2$ ), and the tri-calcic aluminate ( $3 \text{ CaO}$ ,  $\text{Al}_2\text{O}_3$ ). The orthosilicate in the clinker is usually in its  $\beta$  (beta) allotropic form. The  $\text{Fe}_2\text{O}_3$  probably forms tri-calcic ferrite ( $3 \text{ CaO}$ ,  $\text{Fe}_2\text{O}_3$ ). If the cement is overlimed, or underburnt, free CaO may be present; underburnt cement may also contain the compound  $5 \text{ CaO}$ ,  $3 \text{ Al}_2\text{O}_3$ .

**430.** The clinker coming from the kilns is ground to a flour and becomes the main constituent of Portland cement. Ground-up clinker cannot be used alone as hydraulic cement because it sets hard after the addition of water within ten minutes, and therefore could not be handled in commercial construction. In order to make the setting or beginning of hardening of the cement reasonably slow, calcium sulphate ( $\text{CaSO}_4$ ) is added either in the form of plaster of Paris ( $\text{CaSO}_4$ ,  $\frac{1}{2} \text{ H}_2\text{O}$ ) or gypsum rock ( $\text{CaSO}_4$ ,  $2 \text{ H}_2\text{O}$ ). The amount of sulphate added is sufficient to make the weight per cent of  $\text{SO}_3$  in the Portland cement about 1.5. The gypsum (or plaster of Paris) is preferably mixed and ground with the clinker.

Cement should be stored for some time before use. This "seasoning" allows the moisture and  $\text{CO}_2$  of the air to react with free lime (CaO) changing it to hydrate or carbonate of lime, improving the

soundness of the cement. There are probably other changes in the cement during seasoning. It is well known that many cements which are unsound when freshly made become good by storage.

**431.** When cement is mixed with water it makes first a paste or mud. The water becomes noticeably alkaline from the presence of calcium hydrate in solution. Very soon the paste stiffens, becoming semi-plastic. This action is known as "setting." After setting has begun the cement should not be disturbed because if broken and then put together again the bonding on the surface of break is imperfect. For this reason "initial set" should for commercial purposes not take place until a reasonable time has elapsed sufficient for the mixing, transportation and placing of the cement.

Some hours after the initial set the cement suddenly begins to gain in strength or to "harden." The gain of strength is very rapid for about 24 hours after the beginning of hardening and continues at a decreasing rate for about 3 months. Most cements after 3 months of age decrease somewhat in strength over periods of 1 to 3 years, then becoming constant or slightly rising in strength. In defective or "unsound" cements this drop of strength may come sooner and be so serious as to destroy the structure in which the unsound cement is used.

**432.** The chemistry of the setting and hardening of Portland cement has been most thoroughly studied by Messrs. Kline and Phillips of the U. S. Bureau of Standards. Their summary of the actions after addition of water or "gaging" of the cement follows. "On the hydration of cement, the first constituent to react is the aluminate, with the formation of amorphous hydrated tri-calcium aluminate, with or without amorphous hydrated alumina (calcium hydrate going into solution). The sulpho-aluminate crystals are also formed and the low-burnt or finely ground lime forms hydrates. This occurs within a few hours after the cement is gaged. The next compound to hydrate is the tri-calcium silicate. This commences within twenty-four hours and is generally completely hydrated within seven days. Between seven and twenty-eight days, the amorphous aluminate commences to crystallize and the beta orthosilicate, the least reactive compound, begins to hydrate. The twenty-four hour strengths are due mainly to the hydration of the aluminates and of any fine grained, low-burnt

lime present. The large increase in strength between twenty-four hours and seven days is due mainly to the tri-calcium silicate hydration. The increase between seven and twenty-eight days is due to the hydration of the beta orthosilicate. Where there is a decrease in strength during this period it is due to the hydration of very high-burnt free lime as in very high-burnt, high-limed cements, or to the crystallization of the aluminates as in high-alumina cements. The iron compounds in a cement are resistive to hydration. It does not form crystalline hydration products but occurs as a rust-like material.

"The initial set of cement is affected by the action of small amounts of electrolytes in retarding coagulation of the aluminate material. With a limited amount of water, such as used in normal consistency mixes, the aluminates coagulate and separate from supersaturated solutions as amorphous bodies, the rate of coagulation being affected by such small quantities of electrolyte as to nullify the possibility of the reaction being solely a chemical one."

In comment on the above we may note that the iron compounds in the cement are almost certainly hydraulic material of value similar to the aluminates, else the high-iron cements would not have the excellent properties they do. The effect of electrolytes on time of initial set may be either retardation or acceleration, depending upon the nature and concentration of the ions of the electrolyte.  $\text{CaSO}_4$  is a retarder of initial set.  $\text{CaCl}_2$  in small amounts (up to 1 or 2 per cent of the weight of the cement) greatly accelerates the setting and hardening, but in larger amounts loses this action and may ultimately become a retarder. The peculiar and often puzzling action of ground waters containing slight amounts of chemicals in solution, in retarding or accelerating the setting of cement, perhaps is due to this electrolytic action. The bodies called amorphous by Kline and Phillips are the same that have been called "colloids" or "gels" by Michaelis and others. Their action is somewhat similar to that of glue, at first soft and weak and gradually hardening to considerable strength.

The particles of the ground cement are somewhat less than  $\frac{1}{200}$  inch in diameter. When the cement is mixed with water the particles become coated with a water film which reacts with the cement. The first product of the reaction is an amorphous or jelly-like layer surrounding the cement particle. As time goes on, the water penetrates

more deeply, more and more of the material of the cement particle combining with water. The jelly apparently dries out or hardens and crystals may form.

**433.** Portland cement is used mainly in the making of concrete. The greater part of the volume and mass of concrete is rock. The holes or "voids" between pieces of rock are filled with smaller rock or gravel. The voids through the gravel are filled with sand. Lastly the voids between sand particles are filled with cement, and enough excess cement is added to make sure of coating the surfaces of rock, gravel and sand with the cementing material. The cement therefore bonds the whole aggregation into a unit. Too little cement will weaken the concrete by lack of complete bonding, while excess of cement over that needed for complete bonding does not increase the strength of the concrete in proportion to the increased cost and is uneconomical. The amount of cement required depends principally upon the volume of voids to be filled. The voids through the concrete are minimum when the sizes of rock, gravel and sand are graded so that the voids of each are just nicely filled by the next smaller size of material. This principle of grading for minimum voids is commonly known as grading for maximum density. It has been repeatedly demonstrated that this gives the best concrete.

The maximum size of rock pieces depends upon the character of the structure. For massive foundation work where the highest strength is not necessary the rock sizes used are largest. The maximum rock size must not exceed a small fraction of the thickness of the concrete structure. For reinforced concrete the maximum size of stone is often limited to one inch, to make sure of proper packing around the steel. The sand must not be too fine. If the size of sand particles is too near the size of cement particles the cement is unable to coat thoroughly the surfaces of the sand particles. Experience seems to show that sand smaller than 30 mesh, or about 0.02 inch diameter of particles, is undesirable. The sand should preferably be graded in size rather than uniform. Whether rock and sand are crushed or not makes little difference provided the surfaces are clean. Rounded material has somewhat less voids than the angular crushed material. The surfaces of crushed material are fairly clean. The surfaces of uncrushed material are very likely to be covered with films of clay or

vegetable matter which prevent the cement, from contact with the rock or sand and seriously decrease the strength. Washing to remove the clay and vegetable matter must be thoroughly and carefully carried out. Crushed material is generally preferred because it does not need this washing. The chemical nature of both rock and sand must also be considered. Mica is very bad; so also is any material, such as pyrites,  $\text{FeS}_2$ , which may react with the cement in time, causing disintegration.

434. The materials for concrete are usually measured for mixing in parts by volume. The composition of the concrete is stated by the ratio of volumes of cement to sand to rock. Such ratios may vary from  $1 : 1\frac{1}{2} : 3$  to  $1 : 4 : 8$ . The first ratio would go with small-sized rock and a "rich" mixture, and might be used in reinforced concrete of the highest strength. The second ratio is that of a "lean" concrete such as might be used in foundation work. Concrete should not be mixed on these volume ratios blindly. The rock and sand should be analyzed for their percentages of different sizes and the proportions of the two so chosen as to obtain the minimum of voids. In addition one must remember that the finer the rock and sand the greater are the surfaces exposed to be covered with cementing material. Since cement must both fill all remaining voids and coat the surfaces of both sand and rock, more cement is required in proportion to sand and rock as the average rock size decreases.

435. The mixing of concrete is now done almost entirely by machinery. This has been found not only cheaper but better than hand mixing. Very commonly rock, sand, cement and water are put into the mixer together, and the charge is given a certain time of working at a fixed speed. The tendency is to accelerate output to the point of insufficient mixing, making inferior concrete. It is found by microscopic study of concrete that a considerable proportion, 30 per cent or more, of the cement is not hydrated. Unhydrated cement in the concrete is not entirely wasted for the voids must be filled in any case. It is obvious, however, that more complete hydration would make a stronger concrete with the same materials.

The principal reason for lack of hydration of the cement is an effect of the surface tension of the water. The water films tend to pull the cement particles together in groups. These groups of particles

hydrate externally but not internally. Partial remedies may be found in several devices: (1) mixing the sand, cement and rock dry before adding water and mixing again afterward, will distribute the cement better before water touches it; (2) longer and more thorough mixing may mechanically break up the groups of cement particles; (3) anything which will reduce the surface tension of water, such as warming the water, or addition of slight amounts of certain chemicals in the water, will markedly reduce the power of the water films to group the cement particles.

**436.** In the mixing, after the addition of water, air bubbles are unavoidably entrapped in the mortar. If the mixing is very splashy and the concrete is carelessly thrown into place in the forms it is possible to have as much as thirty per cent of the mortar made up of the various sizes of air bubbles. The air bubbles have a tendency to collect along surfaces of gravel or stone, preventing contact of the cementing material to those surfaces. The air bubbles therefore have a very sharp effect in weakening the concrete. Examination of concrete which has been loaded beyond its elastic limit shows a very general tendency for the incipient cracks to start from air bubbles or along surfaces of stone. The only remedy for air bubbles seems to be the preventive method of taking care in the mixing and handling.

**437.** The amount of water used in mixing should bear a definite ratio to the other materials. The immediate function of the water is to make a cementing paste out of the cement, and to coat the surfaces of sand and rock with a film of water. Too little water fails to develop the full strength of the cement paste and makes a concrete so dry that it will not pack itself in the forms. This increases the cost of placing. Sufficient water makes a concrete just on the point of being able to flow in the forms. It requires very little tamping into place. Excess water makes the concrete very fluid, reducing the labor cost of placing, but delays the setting and hardening and probably reduces the final strength. If the molds are not water-tight excess water will wash out cement. The cement cannot absorb beyond a certain amount of water, and the excess must either remain as water in the concrete, or if it escapes leaves voids.

**438.** The rate of hardening of concrete, or the time required to reach strength enough for the new structure to take care of itself, depends



very much on temperature. The lower the temperature the slower is the hardening. Danger comes when freezing of the water may occur. If ice forms from the mixing water in the concrete before the hardening, the concrete is cracked and will not reunite on the cracks when the ice melts and setting or hardening continues. Because of slow hardening and danger of freezing, concrete construction is little done in the winter season. If it must be done then, salt is used to lower the freezing point of the water, or more commonly the rock, sand and water are warmed before mixing and kept warm until hardening is well under way.  $\text{CaCl}_2$  is better than  $\text{NaCl}$ , because it accelerates setting as well as lowers the freezing point of the water.

**439.** Cement and concrete undergo volume changes during hardening. In air the volume change of cement is a contraction amounting to about  $\frac{1}{6}$  of 1 per cent in the first month and ultimately becoming about  $\frac{1}{3}$  of 1 per cent. In water cement expands about  $\frac{1}{20}$  of 1 per cent in the first month and ultimately about  $\frac{1}{8}$  of 1 per cent. The volume changes of concrete are less than those of cement in the volume ratio of cement to sand plus rock in the concrete. Thus a 1 : 3 : 6 concrete would have expansions or contractions about  $\frac{1}{8}$  that of the cement. While these changes appear small, they may set up serious stresses. A 1 : 3 : 6 concrete in air ultimately has a linear shrinkage of  $\frac{1}{3} \times \frac{1}{8}$  or  $\frac{1}{24}$  of 1 per cent. With a modulus of elasticity of 2,000,000 this means a stress of 740 pounds per square inch, which is considerably higher than the tensile strength of the concrete. Necessarily the concrete must be cracked by this shrinkage unless provision was made in laying the concrete by putting in expansion joints. The shrinkage shows up also in surface cracking of the concrete. In smoothing off the surface of the wet concrete a very rich surface layer is formed which in hardening will have much more shrinkage than the leaner concrete beneath. The more rapid hardening of the surface layer due to loss of water by evaporation, further exaggerates the difference between this layer and the rest of the concrete. The shrinkage causes great numbers of minute superficial cracks. These cracks may then become starting points for weathering action by water and frost.

**440.** Concrete is relatively strong in compression and shear and weak in tension. The strengths should be: compression, over 2000 pounds per square inch; shear, over 1000 pounds per square inch;

and tension, over 200 pounds per square inch. The weakness in tension makes it necessary to avoid any serious exposure of the concrete to tension stressing, whether that stressing be primary or secondary. Thus the primary and diagonal tensions in a concrete beam or the hoop tensions in a concrete column are likely to be the dangerous stresses. In order to make up for the deficiency of concrete in tensile strength, steel pieces are buried in the concrete in such positions as to carry the tensions. This combination material, using the concrete for compression stresses and steel for tensions, is called **reinforced concrete**. The reinforcing is possible because the cement adheres to the steel more than strongly enough for the transfer of stresses from one to the other, while the stiffness of the steel is about fifteen times that of the concrete. The two materials deform together under load. A relative stretch of 0.0001 would set up in the steel a stress intensity of  $0.0001 \times 30,000,000 = 3000$  pounds per square inch and in the concrete a stress of  $0.0001 \times 2,000,000 = 200$  pounds per square inch. This shows that a relatively small proportion of steel will take the load away from the concrete. The concrete protects the steel from corrosion.

**441. The testing of cement and concrete differs radically from that of other materials of construction because the properties change with time.** Starting with zero strength when first mixed with water, the strength of cement or concrete rises very rapidly for a few days, then slowly for some months before it is approximately at its final value. There is often a drop of strength after three months to even some years, and this decrease may go so far as to destroy the structure, or the strength may again rise. The testing is usually limited to determinations not taking over one month of time which are used to give some indication of the probable future action of the cement. Where important concrete construction work is under way it is desirable to make up from a part of the concrete compression-test blocks, to be stored under the same weathering conditions as those to which the structure is exposed, and tested at intervals up to an age as great as five years. By this device knowledge will be obtained of the probable conditions of the entire structure and possibly serious accidents may be avoided.

**442.** The setting and hardening of cement or concrete are greatly affected by temperature and somewhat affected by the amount of

water used in mixing. These factors affect short-time tests, such as are commonly used, more than they do long-time tests or the final properties in actual construction. Assuming that the temperature effect can be represented, as in the heat treatment of steels, by a variation of chemical reaction rate with the 20th power of the absolute temperature, the rate of setting and hardening or relative time for reaching a given stage of hardening will vary with temperature as in the following table.

EFFECT OF TEMPERATURE ON SETTING AND HARDENING

Temperature °F.	Relative time.	Relative rate.
0	17.1	0.058
20	7.24	0.138
32	4.41	0.227
40	3.21	0.312
50	2.17	0.461
60	1.47	0.680
70	1.000	1.000
80	0.685	1.46
90	0.468	2.14
100	0.332	3.02
150	0.060	16.7
200	0.0124	80.8
300	0.0074	135.0

The variation of rate of action with temperature shown in this table is at least approximately correct over the temperature range in which testing has commonly been done (32° to 100° F.). **Because this effect of temperature is so large it is necessary to adopt a standard temperature for cement testing in order to make results obtained in different laboratories or at different times comparable with one another. The standard temperature adopted is 70° F.** In the above table the rate of action was made unity at this standard temperature.

**443.** The effect of increasing the amount of water used in mixing is to postpone the initial set of the cement very decidedly and to slow up the hardening of the cement to an appreciable extent. Within the ordinary ratios of weight of water to weight of cement, say 20 to 30 per cent, the time of initial set varies approximately with the 4th power of the percentage of water. This fact is of considerable importance in the commercial handling of concrete because it gives the

## INSTRUCTION

so that the concrete can be tested of cement it becomes water used in mixing; other various ages would not be and be remembered that it is the cement so that it cannot

Amount of water used in mixing determination for each cement consistency" ratio of weight of weight of cement. A very of finding normal consistency thoroughly and rapidly with known weights of cement and make up from the paste a ball of 4 inches in diameter. This ball is dropped to the testing table from a height of sixteen inches to two feet. If it breaks in pieces it is too dry; if it is more than half its diameter it is too wet. This simple test is sharper than the cone test would indicate, for the difference in amount of water between a mixture that is too dry and one too wet is not large. The consistency test must be done at a constant temperature. The elapsed time between the adding of water to cement and the test should not exceed five minutes and preferably should be less than three minutes, though time must be allowed for thorough working. The table on which mixing is carried out should be impervious to water. If it is not, it should be wet with water before use.

The normal consistency is by the test of a paste of known weights of cement and water mixed together then quickly and thoroughly

pushed into the rubber ring *G* of the tester and struck off flush with the edges. The one-centimeter diameter plunger *C* is brought to contact with the upper surface of the cement paste, the scale reading taken, and the plunger released. The plunger should have a jump penetration of one centimeter if the paste is of normal consistency. This test is even sharper than the other and agrees fairly well with it. The same precautions as to time and temperature must be observed here as in the ball test.

**445.** Normal consistency varies from one cement to another with differences in chemical composition, but is influenced most largely by the fineness of the cement. The finer the cement the more water is required for normal consistency. As the object of the normal consistency determination is to put the time of set tests and strength tests on comparable bases, so far as amount of water used in mixing is concerned, it would seem that the desired object could be secured quite as well by an arbitrary requirement that the ratio of water to cement should in all cases be, say, 25 per cent. This is rather more water than the normal consistency test usually requires, but not a great enough excess to materially alter times of set or strengths at small ages. It will give higher strengths at greater ages than the present normal consistency does. It would favor the finely-ground cements in contrast to the coarsely-ground cements, yet in none would it make a paste too wet to handle well. It is certainly open to question whether the use of a certain definite percentage of water for all cements would not be just as fair a basis of comparison as the present normal consistency test. It would greatly facilitate testing by doing away with the normal consistency determination, which is unsatisfactory both in description and operation.

**446.** When tests are to be made upon mortars it is necessary to have a **standard sand** because the strength of a mortar is as much a function of the sand used as of the cement. This standard sand is arbitrarily chosen and is intended to be used in all testing laboratories whenever the bonding value of a cement is determined in mortar tests. The standard sand is a clean quartz sand, water worn, graded to pass a 20-mesh and not to pass a 30-mesh sieve.

**447.** The most important test of a cement is probably that for **soundness**. From a normal consistency paste of cement and water

("neat" cement) pats are made up on glass plates. The pats are about 3 inches in diameter, about  $\frac{1}{2}$  inch thick at center, tapering to thin edges. Their surfaces should be smoothed with a trowel. These pats are first hardened for 24 hours at 70° F. During this hardening they should be protected against evaporation of the water used in mixing, either by keeping them covered with a moist cloth or by storage in a "wet cabinet." After this hardening some of the pats are kept in air at 70° F. for 28 days, others in water at 70° F. for 28 days, while a third set are cooked in steam over boiling water or actually in the boiling water for 3 to 5 hours. The last is called an "accelerated" test because the high temperature causes much more rapid action. The object of the soundness test is to find out if there is any chemical in the cement, free CaO or MgO for instance, which will cause excessive or irregular volume changes in the cement after hardening. The soundness test is sometimes called the test of constancy of volume. The hydration of free CaO or MgO after the cement has hardened cracks the cement. Unsound cement will develop radial cracks starting from the edges of the pats. Such cement would be rejected no matter what the other tests might show. There may be harmless shrinkage cracks due to the normal shrinkage which all cement undergoes in air hardening, but these cracks parallel the edges of the pats or occur at the center and not at the edges. Warping of the pats away from the glass plates does not prove that the cement is bad, but raises the question so that further testing is necessary. It should be remembered that many cements are slightly unsound when freshly made and that they will become sound with storage or seasoning. Unsound cements are too dangerous to use in any structure of importance, because concrete made with them will in time disintegrate spontaneously.

448. In general the boiling test, besides being immensely quicker than the 28-day test in air or water, is a harder test to pass, especially if boiling under water for 5 hours is used. It is not probable that any unsound cement can stand the latter test. Following out this line of reasoning a still more severe test is now being used for soundness, known as the "autoclave" test. In this the boiling is performed in a closed vessel in which considerable steam pressure can be had and the temperature can be carried up to 300° or 400° F. It has been

shown recently by the U. S. Bureau of Standards that the chemical reactions between cement and water are of the same kind at these high temperatures as at ordinary temperatures. The difference lies in the speed of reactions. The table in Section 442 shows that at 300° F. the rate of action should be about 135 times that at 70° F. Therefore 5 hours at 300° F. is equivalent to 28 days at 70° F. The hydration of free lime, however, is brought out even more sharply than these figures would indicate.

**449.** Two "times of set" are reported in a cement test. The first is called the "initial" set, the second the "final" set. They represent definite stages of the setting and hardening reactions. The times of set are measured from the time of adding water in mixing the cement paste to the time when each definite condition of strength first occurs. The initial set is intended to be that stage of setting after which, if the cement is broken, it will not thoroughly reunite along the surfaces of break. Cement or concrete should be in place in construction work before the time of initial set. In order to have a reasonable opportunity for mixing, transporting and placing the concrete the specification has been made that the time of initial set must not be less than  $\frac{1}{2}$  hour. Time of initial set is controlled mostly by the amount of gypsum added in making the cement. It occasionally happens that in the storage of cement the gypsum loses its effectiveness and the cement becomes quick setting. Most American Portland cement has an initial set time of from 2 to 3 hours. In the commercial making of concrete the water exceeds the normal consistency amount, further delaying the time of set, and temperature effects must also be expected.

The final set is intended to mark the beginning of appreciable strength of the cement. For about 24 hours after the final set the strength rises very rapidly. A cement which takes too long for its final set is slow hardening, gains strength slowly, and may delay construction work by requiring the forms to be left on much longer than usual. Forms are usually left on from 24 to 48 hours on construction not reinforced, and from 1 to 4 weeks on reinforced construction, where the material must have nearly full strength before being compelled to take care of itself. The limitation put by specification on time of final set is that it should not exceed 10 hours. The limitation ought not to be applied to the time of final set, but rather to the dif-

ference of time between initial and final sets. Changing the amount of water in mixing, for instance, makes a parallel shift of both initial and final set times, not greatly affecting the difference of time between the two. Since the strength at initial set and at final set has definite values, the quotient of this strength difference by the difference of time of the two sets establishes the slope of the strength-time curve or the rate of gain of strength with time as the cement hardens, or we may say that the rate of gain of strength varies inversely as the difference of time between initial and final sets.

**450.** There are two methods in common use for measuring the times of set. One uses the Vicat tester, Fig. 179; the other the "Gilmore needles." For the test with the Vicat apparatus a neat cement paste of normal consistency is made and placed in the ring of the tester as in the normal consistency test. A damp cloth should be put over the cement to prevent evaporation of mixing water from the upper surface. Usually this precaution is neglected. The millimeter diameter needle of the tester is from time to time brought to the upper surface of the cement in the ring and released. At first the needle will go clear through the cement to the bottom plate. As the cement sets and stiffens the penetration of the needle decreases. The depth of the ring is 40 millimeters. When the needle first fails to penetrate the last 5 millimeters from the bottom, initial set has occurred. When the needle first ceases to have appreciable penetration from the upper surface, though still marking the surface, final set has occurred or hardening has begun.

**451.** There are two Gilmore needles. One, for the initial set, has a brass ball weighing  $\frac{1}{4}$  of a pound used as a weight on a cylindrical steel needle with an end diameter of  $\frac{1}{16}$  of an inch. The other, used for the final set, has a ball of 1 pound weight on a steel needle of  $\frac{1}{8}$  inch diameter. The cement paste for use with the Gilmore needles is made up on glass plates as for the soundness test. A damp cloth should be put over the pats to prevent evaporation of the mixing water. For test, the point of one of the needles is brought flush with the surface of the cement near the center of the pat and the weight of the needle gradually released. The times of set are taken when the cement is first strong enough to hold the needle from appreciable penetration though still marking the surface.



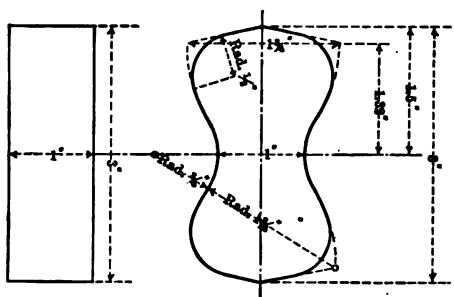
**452.** The neglect of the precaution to prevent evaporation of mixing water from the surface affects both the Vicat and Gilmore tests of times of set. It shortens the reported times because cement sets more quickly with less water. The Vicat initial set time is not greatly affected, because of the great depth of cement through which the needle is working. The Gilmore test will be affected both in initial and final sets because they deal exclusively with the surface of the pats. The general criticism runs against the Vicat final set test and both the Gilmore tests that dealing with the surface of the cement involves the maximum possible errors in the making of these tests. It would probably be better to standardize all time of set tests with appreciable penetrations by the needles. The penetration-time curves of a weighted needle are at first flat, then steep and again flat. Sharp measurements of setting time can be had only on the steeper portions of such curves, which means that much more precision would be obtained if surface measurements were eliminated.

**453.** Fineness of grinding is important to cement both before and after burning. Fine grinding before burning insures thorough chemical combination in the burning, without which the cement may be weak or unsound or both. Fine grinding after burning is needed to increase the ability of the cement to react with water, and to enable the cement particles to coat the surfaces of sand in mortar. The finer the cement the stronger is the mortar produced with a given sand. Increased fineness makes a cement quicker in setting and hardening. Probably it may be assumed that a cement finely ground in its finished form also received fine grinding before burning and therefore a good chance of proper burning.

No very satisfactory method of measuring the fineness of cement has yet been found. The present method is to measure the percentage by weight of the cement capable of passing various sizes of sieves. The sieves are made of brass wire of standardized sizes of wire and number of wires or meshes per linear inch. The sizes used are 50, 100 and 200 mesh. A good cement should not leave more than a few tenths of one per cent of its original weight on a 50-mesh sieve, nor more than 8 per cent on a 100-mesh, nor 25 per cent on the 200-mesh, according to present specifications. The difficulty with this system is that we do not know the real fineness of the material

passing the 200-mesh sieve. This material happens to be the greater and by far the more valuable part of the cement. A known amount of cement is placed in the sieves, which are nested in order of fineness, and thoroughly shaken through. The end point for each sieve is considered to be reached when the amount passing in one minute of shaking goes below  $\frac{1}{10}$  of 1 per cent of the original weight.

**454. Strength tests of cement are not made to find out how strong a structure made with the cement will be, but to get data confirming and amplifying the information obtained from the soundness, times of set and fineness tests.** The strength of a concrete used in an actual structure depends upon the stone, the sand, the cement, the water used in mixing and the general method of mixing and placing. A concrete is always loaded in compression except where reinforcing is used. Strength tests of cements (not of concrete) are usually made in tension. The argument for using the tension test was that the lower strengths allowed the use of small and simple machinery.



DETAILS OF STANDARD TEST  
PIECE (BRIQUET).

FIG. 180.

Experience has now shown that the variability of the tension tests on pieces supposedly identical is very much wider than the similar variability in compression tests. It is probable that in time the tension testing will be dropped and the standard tests will be made in compression. It was formerly thought that the ratio between compression and tension strengths was a definite function of the age of the cement, the same for all cements. This is now known to be far from the truth. Compression results cannot be closely predicted from the tension results. The ratio of compression to tension strengths in

neat cement pieces varies with the fineness of the cement; in mortars it varies both with cement and sand; concrete introduces still more variables. The ratio is of the general magnitude of 10 : 1, but may be anywhere from 5 : 1 to 20 : 1.

**455.** The standard tension tests are made upon pieces of the form and dimensions shown in Fig. 180. Tests are made on neat cement and on a 1 : 3 mortar made of 1 part by weight of cement and 3 parts by weight of standard sand. Neat cement "briquets" are tested at ages of 1, 7 and 28 days. The mortar briquets are tested at 7 and 28 days. The minimum strengths which must be shown in these tests are specified as follows:

MINIMUM TENSILE LOADS OF STANDARD PIECES

Material.	Age.	Load, lbs.
Neat cement	1 day	175
Neat cement	7 days	500
Neat cement	28 days	600
1 : 3 mortar	7 days	200
1 : 3 mortar	28 days	275

**456.** The neat cement pieces are made from a mix of normal consistency, thoroughly packed into the briquet molds and struck off flush. The molds should be clean and very slightly greased to make the briquets come out easily. The 1 : 3 mortar has the cement and sand thoroughly mixed dry, then the total water is added and mixed in. The water required for the mortar has a weight equal to the sum of normal consistency percentage of weight of cement plus 5 per cent of the weight of the sand.\* Both kinds of test pieces must be protected from evaporation of water, either by covering with damp cloths or by putting into a wet cabinet. After hardening for 24 hours the pieces are either tested for a 1-day test or are put under water to remain there until tested. The molds are taken off at the end of the 24 hours in the wet cabinet.

**457.** In testing, precaution must be used that the test pieces are placed squarely in the jaws of the testing machine. Eccentric loading will seriously decrease the breaking load. The rate of loading must

\* This is not the standard rule of the A. S. T. M., but is a rational rule giving about the same result.

be standardized. Rapid application of load gives higher strength than slow application, so much so that failure to observe the standard rate of loading spoils the results. The standard loading rate is 600 pounds per minute.

**458.** Although the minimum cross-section area of the standard tension test pieces is just one square inch, the breaking load is not the tensile strength of the piece in pounds per square inch. The stress intensity is far from uniform across the neck of the test piece. The maximum stresses at the edges of the test section are more than twice the minimum stress at the center of the test section and a little over 1.5 times the average stress, which is the one ordinarily computed. The true tensile strength of cement or mortar may be taken as 1.5 times the recorded strength of the standard test pieces.

Almost equally fallacious is the assumption that load over area, for the common cubic form of compression test piece, gives the compression strength in pounds per square inch. The true compression strength is about 85 per cent of that found from a cubic test piece. The test piece should be a cylinder or prism of height at least 2.5 times its diameter or least thickness. For the reasons for this see Section 150.

**459.** Fig. 181 shows the curves of tension strengths versus time for sound and unsound cements. It will be seen that in neat cements the early strength of the unsound material is higher than that of the sound cement, and the relation reverses as age increases. In the 1 : 3 mortar the unsound cement is at all ages weaker than the sound. To these curves might be added others. Underlimed cements are slow hardening. While their final strengths are good enough, the rise of strength is slow and construction work would be very much delayed. Underburnt cements may be both weak and unsound. The unsound curve shown in Fig. 181 comes from overliming. Fine grinding of the cement lowers the strength of neat cement pieces in short-time tests, but raises the strength of the mortar pieces. The contradiction between the neat cement and the mortar tests as to the results of unsoundness or fineness, makes it necessary to be exceedingly careful in interpreting the results of the standard short-time tension tests and to make such interpretation only in connection with the soundness, times of set and fineness tests. Judgment should be

based much more upon the mortar tests than upon the neat cement tests.

**460.** A test which has now been discontinued is that for specific gravity of the cement. This test was originally considered to be a test for adulteration. The adulterants possible lie so near in specific gravity to that of cement that the test is not delicate. The specific gravity of Portland cement freshly made is between 3.1 and 3.2; by seasoning it decreases to about 3.1. Adulteration would be discovered by the tests for soundness, times of set and strength long before it would be shown by the test for specific gravity.

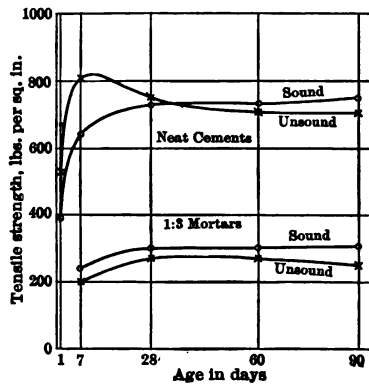


FIG. 181.

Tensile Strength-Time Curves of Sound and Unsound Cements. Data of Mr. W. P. Taylor, *Trans. A.S.T.M.* vol. 3, 1903.

**461.** Adulteration may also be found by chemical analysis. In the cement itself limitations are set by specification upon the amounts of certain components.  $MgO$  may not exceed 4 per cent, nor  $SO_3$  1.75 per cent. This limitation is set because of fear that too much  $MgO$  or  $SO_3$  may make the cement unsound. Chemical analysis of a cement is not generally made in commercial testing.

**462.** Color of a Portland cement has nothing to do with its quality. Cements containing no  $Fe_2O_3$  are nearly white. With  $Fe_2O_3$  they have brownish shades, more pronounced as the  $Fe_2O_3$  increases. The properties seem to be equally good whatever the relative proportions of  $Fe_2O_3$  and  $Al_2O_3$ , except that the resistance to chemical action, as by sea water, is better with the high-iron cements.



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